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A PRELIMINARY INVESTIGATION OF THE CORROSION AND STRESS-CORROSI--ETC(U)

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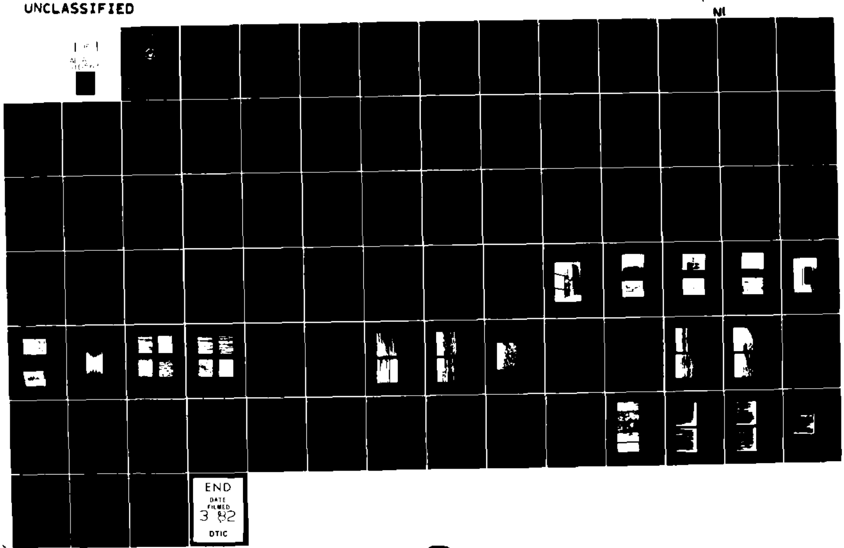

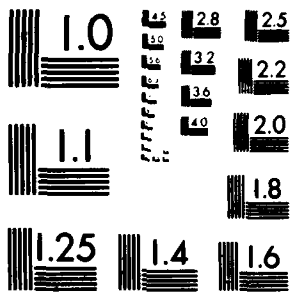

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THESIS

A PRELIMINARY INVESTIGATION OF THE CORROSION AND STRESS-CORROSION SUSCEPTIBILITY OF THERMOMECHANICALLY PROCESSED HIGH MAGNESIUM ALUMINUM MAGNESIUM ALLOYS

by

Larry Edward Beberdick

September 1981

Thesis Advisor: T. R. McNelley

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REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER	2. GOVT ACCESSION NO. AD-A110	3. RECIPIENT'S CATALOG NUMBER 563
4. TITLE (and Subtitle) A Preliminary Investigation of the Corrosion and Stress-Corrosion Susceptibility of Thermomechanically Processed High Magnesium, Aluminum Magnesium Alloys		5. TYPE OF REPORT & PERIOD COVERED Master's Thesis September 1981
		6. PERFORMING ORG. REPORT NUMBER
7. AUTHOR(s) Larry Edward Beberdick		8. CONTRACT OR GRANT NUMBER(s)
9. PERFORMING ORGANIZATION NAME AND ADDRESS Naval Postgraduate School Monterey, California 93940		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS
11. CONTROLLING OFFICE NAME AND ADDRESS Naval Postgraduate School Monterey, California 93940		12. REPORT DATE September 1981
		13. NUMBER OF PAGES 84
14. MONITORING AGENCY NAME & ADDRESS(if different from Controlling Office)		15. SECURITY CLASS. (of this report) Unclassified
		15a. DECLASSIFICATION/DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (of this Report)

Approved for public release; distribution unlimited.

17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)

18. SUPPLEMENTARY NOTES

19. KEY WORDS (Continue on reverse side if necessary and identify by block number)

Thermomechanically Processed, High Strength, Stress Corrosion, Aluminum, Corrosion, Aluminum Magnesium, Marine

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

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or Cu and Mn to a 10% Mg alloy raises strength, homogenizes the microstructure and reduces the tendency of such an alloy to exhibit intergranular cracking and exfoliation, especially in a sensitized condition. Results of accelerated general corrosion testing and marine exposure both indicate that binary 8% Mg and 10% Mg alloys are highly resistant to corrosion. Alloying with Cu or Cu and Mn accelerates weight loss but to a lesser degree than observed for 7075-T6.

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A Preliminary Investigation of the Corrosion and Stress-Corrosion Susceptibility of Thermo-mechanically Processed High Magnesium, Aluminum Magnesium Alloys

by

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Submitted in partial fulfillment of the
requirements of the degree of

MASTER OF SCIENCE IN MECHANICAL ENGINEERING

from the

NAVAL POSTGRADUATE SCHOOL
September 1981

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ABSTRACT

The stress corrosion cracking susceptibility and general corrosion characteristics of four thermomechanically processed high-Magnesium, Aluminum-Magnesium alloys were evaluated and compared to those of 7076-T6. Results obtained from stress-corrosion testing and from tension testing after stress-corrosion exposure indicate that these 8-10% Mg alloys are less susceptible to stress-corrosion cracking than 7075-T6. The addition of Cu or Cu and Mn to a 10% Mg alloy raises strength, homogenizes the microstructure and reduces the tendency of such an alloy to exhibit intergranular cracking and exfoliation, especially in a sensitized condition. Results of accelerated general corrosion testing and marine exposure both indicate that binary 8% Mg and 10% Mg alloys are highly resistant to corrosion. Alloying with Cu or Cu and Mn accelerates weight loss but to a lesser degree than observed for 7075-T6.

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ACKNOWLEDGEMENT

It is appropriate at this time to express my most sincere appreciation to those individuals who have provided help and guidance, and especially those who supported me during this ordeal. I want to thank Mr. Richard Schmidt of Naval Air Systems Command for his support in this research. I want to thank all of the shop and lab personnel who worked in developing the test apparatus. I would also like to thank Dr. J. Perkins for being my mentor in corrosion matters, and Dr. T. McNelley my mentor in microstructural matters, as well as being my advisor. The largest assist came from fellow thesis students who were ever present and acted as sounding boards for observational ideas. My most sincere and heart felt appreciation go to my family, whom I neglected during these last long months, for their constant encouragement, and prayers. Without my wife Jan's loving support, and my children's (Christine and Nathan) unwarranted patience, this document would never have been produced.

I. INTRODUCTION

A. BACKGROUND

The alloys which are the subject of this research are essentially higher Magnesium content alloys than the 5XXX series Aluminum-Magnesium alloys currently in use. They are wrought alloys, and derive their strength through intermediate temperature thermomechanical processing. Wrought Aluminum base alloys are generally classed as either heat treatable or non-heat treatable. The three alloy systems from which the heat treatable alloys are derived are the Aluminum-Magnesium-Silicon system (6XXX), the Aluminum-Copper system (2XXX) and the Aluminum-Zinc-Magnesium system (7XXX). Strength generally increases in the order given, the 7XXX alloys being presently the highest strength Aluminum alloys in commercial use. The non-heat treatable alloys are based on essentially unalloyed Aluminum (1XXX), the Aluminum-Manganese system (3XXX) or the Aluminum-Magnesium (Al-Mg) system (5XXX). In general these non-heat treatable alloys derive their strength from solid-solution hardening and possibly strain hardening.

Conventional wrought 5XXX alloys contain up to 6 percent Mg and are generally considered to be low to medium strength alloys possessing good corrosion resistance, fatigue resistance, ductility and weldability. Their strength, however, is

substantially lower than the high strength alloys (e.g. 7XXX alloys). The strength of such 5XXX alloys can be raised by adding more Mg; however, using conventional processing practice, hot working possibly followed by cold working and annealing, numerous problems arise, including poor resistance to corrosion and stress corrosion and difficulties in cold working.

Conventional high strength alloys, such as the 7XXX alloys, offer high strength but also often suffer from one or more of such problems as poor fatigue resistance and fracture toughness and poor stress corrosion characteristics. Efforts to improve performance in one area (e.g. corrosion resistance) often lead to a sacrifice in another area (e.g. strength). These high Mg, Al-Mg alloys would be similarly subject to limitations and trade-offs and should be viewed as complimenting rather than replacing existing alloys.

B. PREVIOUS WORK

The thermomechanical processing (TMP) method used in this research was developed and characterized in this laboratory as discussed by Johnson [Ref. 1]. That work [Ref. 1] considered both binary Al-Mg alloys and Al-Mg alloys modified with intentional additions of Cu, Mn and Cu together with Mn. Figure 1 is a partial Al-Mg phase diagram illustrating the ranges of Mg content and temperature of interest in this processing method and Figure 2 is a schematic of the essential TMP.

The important features of this processing method are: 1) solution treatment and hot working to dissolve all soluble constituents and homogenize the microstructure; 2) quenching to retain Mg in solution; and 3) reheating to a temperature below the solvus for Mg and extensive warm working to develop a dislocation substructure and stable dispersion of the intermetallic beta ($Al_8 Mg_5$) phase which precipitates under such conditions. The warm rolling below the solvus temperature typically about 300°C for Mg in the alloy results in a homogenous dispersion of equiaxed beta particles in the microstructure. Depending on the details of the TMP and alloy, yield strengths vary from 50 Ksi (345 MPa) to 90 Ksi (620 MPa), ultimate tensile strengths vary from 65 Ksi (450 MPa) to 100 Ksi (690 MPa) and ductility from 18% elongation down to 5% elongation.

In a recent study, Cadwell [Ref. 2] evaluated the fatigue characteristics of two alloys as a function of processing history. His study was especially concerned with the quenching step between the solution treatment and the hot working step and the final warm rolling of the alloy. The significant observation was that a slower quench (oil as opposed to water) leads to improved fatigue behavior under high cycle fatigue (HCF) conditions.

The S-N curves for a 10% Mg alloy, oil quenched between solution treatment and warm working, suggested a fatigue limit at a stress amplitude above 30 Ksi, corresponding to a fatigue-

strength to ultimate-strength ratio of 0.47. In contrast, the alloy subjected to a water quench before warm rolling appeared to exhibit a lower fatigue strength and less tendency for flattening of the S-N surve under similar HCF conditions. This observation was suggested to be the result of precipitation of some relatively course beta particles in the slower quench and these served to homogenize slip, postponing fatigue crack initiation. The fatigue resistance of these alloys, then may be considered as excellent although sensitive to microstructural variations. This current work follows that of Johnson [Ref. 1], as did that of Cadwell [Ref. 2], but with regard to stress corrosion cracking (SCC) susceptibility as the object of concern.

C. THIS WORK

It has been suggested [Refs. 3,4,5,6] that preferential precipitation along grain boundaries in the form of a continuous precipitate film causes susceptibility to SCC in Al-Mg alloys. Consequently, two ways have been proposed to produce SCC resistant microstructures: one, keep the grain boundaries free of the continuous film of precipitates, or two, stimulate precipitation throughout the grains. The first approach is successful if the Mg content of the alloy is below 3%. In the higher Mg, Al-Mg alloys a network of continuous grain boundary precipitates can develop during precipitation at ambient temperatures. To simulate long term ambient aging, a sensitizing

anneal, consisting of heating to 100°C for 7 days, is often used. By comparison, it is not known exactly how fast continuous networks develop at ambient temperatures, but the probable time is as great as 50 years [Ref. 3].

One way of achieving a homogenization of the precipitates is the TMP reported by Johnson [Ref. 1]. Hence a study into the corrosion and stress corrosion characteristics of these warm-worked high Mg alloys is indicated.

This work is intended to be a preliminary study of the general corrosion and stress corrosion susceptibility of a series of warm-worked Al-Mg alloys. Specific areas investigated are: the SCC susceptibility of these alloys in the as-rolled condition in comparison to 7075-T6 alloy; the effect of the sensitizing anneal on the mechanical properties of these Al-Mg alloys; the changes in the SCC susceptibility resulting from the sensitizing anneal; and finally, how these high strength Al-Mg alloys rank among themselves and how they compare to the 7075-T6 alloy in a general corrosion environment.

II. EXPERIMENTAL PROCEDURE

A. PROCESSING AND SAMPLE PREPARATION

Four Al-Mg alloys were studied in this research, their compositions are given in Table I. The alloys were obtained as direct-chill castings from ALCOA, Inc., and are from the same series used by Johnson [Ref. 1], who gave the details of the original manufacture. The processing of these materials was identical to that of Johnson [Ref. 1]. The alloys not containing Mn were solution treated at 440°C for 9 hours, forged, reheated to 440°C and then quenched. They were then reheated to 300°C and warm rolled from approximately 1.0 inch (25 mm) thickness to a final thickness of 0.125 inch (3.2 mm). The material containing Mn was given a second solution treatment prior to warm rolling. This second treatment was at 490°C for 3 hours and was intended to dissolve a Mn-containing phase not taken into solution during the initial 440°C treatment. The 7075-T6 material was obtained as 0.10 inch (2.5 mm) thick sheet.

All subsequent test coupons were prepared from the as-rolled or as-received sheet. These coupons were 3.0 in. x 0.0625 in. x 0.05 in. (76.2 mm x 15.9 mm x 1.3 mm) in size. Due to the limitations imposed by the processing equipment, all samples were prepared with the long dimension parallel to the rolling direction.

Except as noted, all samples were buffed with 600 grit paper and then polished with jeweler's rough. After polishing, the samples were wiped clean with acetone, ultrasonically cleaned in ethanol and then dried in warm air to ensure a smooth, clean grease-free surface. In all cases the samples were subject to the corrosive or stress corrosive environment within an hour of final preparation.

B. ANNEALING FOR SCC SENSITIZATION

Conventionally processed Al-Mg alloys with more than 3% Mg are presumed to be susceptable to aging [Ref. 8]. In order to determine if aging was a problem in these warm-worked high-magnesium alloys, a series of machined test coupons for each of the warm-rolled alloys were given a sensitizing anneal [Ref. 9]. The 7075-T6 alloy was not annealed as the T6 temper is already the most susceptible to SCC [Ref. 7]. The coupons were annealed in a laboratory oven at $100^{o}C$ for seven days (168 hrs). This anneal resulted in some oxidation of the machined surface. One-half of the annealed samples for each alloy were then tested, as described later, with this oxide intact and the other half were polished as above, prior to the SCC exposure.

C. TENSION TESTING

Stress-strain testing was conducted on test coupons representing several conditions examined in this research. Coupons were tested without machining of a reduced gage section, both

for the as-rolled and for materials following SCC exposure. This was done to avoid removing portions of the sample which may have undergone degradation. Samples, then, which failed within the grips of the test machine were disregarded. The tension tests were accomplished using an Instron Model TT-1 test machine set at a crosshead speed of 0.2 in/min (5.1 mm/min) for all tests.

D. STRESS CORROSION TESTS

Stress corrosion testing was accomplished by alternate immersion in a 3.5% NaCl solution. The test cycle was 10 minutes immersed followed by 50 minutes air drying [Refs. 10,11 and 12]. The test chamber, shown in Figure 3, was constructed of Marine-grade plywood and painted with a water-sealing paint. Tank dimensions were 30 inches x 12.5 inches x 50 inches (762 mm x 317 mm x 1270 mm). The salt water in the tank was mainained at a depth of 3.5 - 4.0 inches (89-102 mm), i.e. approximately 20-24 gallons (76-91 l) of liquid.

The sample holder rack (Figure 4) was constructed of Plexiglas and Lexan, and joining was accomplished using epoxy cement and stainless steel screws. A silicone-adhesive caulking compound was used to isolate the screws from the environment. This rack was attached to a pneumatic actuator. This actuator itself was attached to the rear of the tank and, with appropriate timing controls, provided the up and down motion of the rack to accomplish the alternate immersion of the test samples.

Test coupons were stressed in either of two ways. One was a guided U-bend and the other was a three-point controlled bending of the coupon. The U-bend samples (Fig. 5) were formed into a "U" configuration with the procedure and apparatus shown in Figure 6 following the method outlined in Ref. 13. These samples were intended to provide a qualitative comparison of the alloys in the stress corrosion environment. The stress axis is the longitudinal axis of the material. Three-point loaded, bent-beam samples are shown in Figure 7. Stress levels employed were 0.65, 0.8 and 0.95 of the ultimate tensile strength of the material, and were intended to provide a quantative measure of the stress corrosion susceptibility of the alloys tested. The desired stress level was obtained by measuring the vertical deflection of the center of the three-point loaded, bent-beam as shown in Figure 8 and then converting the deflection to stress [Ref. 14].

Three stress corrosion tests were run, each for 740 hours or until failure occurred, at which time only the failed sample was removed. The first set of SCC samples exposed, after polishing, were in the as-rolled/as-received condition. The second set of SCC samples exposed, also after polishing, were in the annealed condition. The third run was done without removing the protective oxide built up during the seven-day anneal. These samples were, however, subjected to the same acetone wipe and ethanol ultrasonic bath to ensure that the only variable in this test was the presence of an oxide layer.

E. GENERAL CORROSION TESTING

The general corrosion test was conducted using a 3.5% NaCl spray following the procedures given in Refs. 15 and 16. The chamber used in this test was contained within the tank used for the SCC tests. However, the spray chamber was isolated from the alternate immersion test by Plexiglas baffel plates. Plexiglas sheets also covered the top of the spray chamber. Figure 3 also shows the 2.0 inch (51 mm) diameter exhaust ducts placed at the rear of each spray chamber to prevent an excessively humid environment in the room in which the apparatus was operating. Within each chamber a 20 inch (510 mm) diameter stainless steel rim rotated at 1/3 rpm in a horizontal plane. The rim was painted with water sealing paint to isolate it from the samples and salt spray. Test samples were cleaned as before and mounted on Lexan holders using a plastic screw (Fig. 9). These holders permitted insulation between adjacent samples and allowed easy access without interrupting the test. Figure 10 shows a rim assembly with a sample and holder installed. Each chamber contained a glass spray atomizer (Fig. 11), which was forced-fed with 3.5% NaCl solution at 13 ml/min. Air was also supplied at a pressure of 5 psig. A single stream of solution was supplied through the center of the nozzle while air was injected in a conical pattern to mix with the solution, forming a fine mist through which the samples rotated. Figure 12 shows a spray nozzle

installation with its associated tubing for air and water. The general-corrosion samples were removed after 5, 10, 50, 100, 500, and 1000 hour points for weight loss measurements.

Additionally, three samples of each alloy were cleaned as before and subject to 1500 hours of actual marine exposure on the Naval Postgraduate School research ship R/V ACANIA. Figure 13 shows the samples and mount before installation aboard the R/V ACANIA.

F. POST-EXPOSURE TESTING

Upon completion of testing all samples were immediately rinsed in warm tap water and scrubbed with a soft bristle brush to remove the salt deposits. They were then immersed for three minutes in a 70% Nitric acid solution to remove corrosion products. The general corrosion samples were then weighted to determine weight loss during exposure. The 1000-hour general corrosion samples were then sectioned for metallographic examination.

The tensile properties of the bent-beam coupons which had not failed in the stress corrosion test were obtained at the conclusion of the stress corrosion exposure. This testing was especially important in this research and was used to provide a ranking of these materials based on degradation of properties resulting from the stress corrosion exposure. This method of evaluation of Al alloys after exposure to a stress corrosive environment was suggested by Budd and Booth [Ref. 17] who conducted electro-mechanical stress corrosion tests on

various Aluminum alloys, and determined that the mechanical properties retained after stress corrosion exposure were indicative of the relative SCC susceptibility of the alloys tested. Such rankings obtained in this research cannot be used to predict actual service lives as appropriate long-term testing was not undertaken here. Nonetheless, such rankings would suggest the relative lifetimes of different materials based on the degree of degradation observed after stress-corrosion exposure.

III. RESULTS AND DISCUSSION

A. THE AS-ROLLED AND AS-RECEIVED MATERIALS

The microstructures and mechanical properties of the warm-rolled alloys processed for this investigation are similar to those reported by Johnson [Ref. 1] in his study of this thermomechanical processing method. Figure 14 shows micrographs of longitudinal sections for the four warm-rolled alloys investigated. The 8% Mg and 10% Mg alloys both exhibit highly elongated grains with some precipitation of the beta (β) phase in grain boundaries. Fewer and smaller β particles are present in the 8% Mg alloy (Fig. 14a) when compared to the 10% Mg alloy (Fig. 14b). The 8% Mg alloy exhibits generally less precipitation within grains whereas the 10% Mg alloy appears to have precipitation within grains and concentrated along slip bands. Cadwell [Ref. 2], in his study of the fatigue characteristics, also examined this 10% Mg alloy by transmission electron microscopy. While still tentative, the results obtained are consistent with the results of this investigation, i.e. that precipitation occurs on slip bands in the 10% Mg alloy. Cadwell [Ref. 2] also observed a fine (0.3 - 0.5 μm) dislocation cell structure in regions distant from the precipitated β particles.

The addition of 0.4% Cu to a 10% Mg alloy results in a more homogenous microstructure after warm rolling (Fig. 14c).

The β phase is still evident in grain boundaries but precipitation is more uniform within the grains, with much less tendency to concentrate in slip bands. The further addition of 0.5% Mn to a 10% Mg - 0.4% Cu alloy results in a still greater degree of homogenization as shown in Figure 14d. There is now no optical microscopy evidence of either grain boundary or slip band preference for precipitation. The size of the β particles in the Cu and the Cu - Mn containing alloys is about the same as in the 10% Mg binary alloy, suggesting a homogenizing, as opposed to refining, effect of these additions. The improved homogeneity evident in the alloy containing both Cu and Mn may not be attributable to alloying effects alone. An additional 3 hour solution treatment of this alloy was necessary at 490°C to dissolve a Mn - containing phase present in the cast condition. This increased solution treatment temperature would contribute as well to a more homogeneous precipitation in subsequent rolling.

The results of tension testing of the 7075-T6 alloy and the four Al - Mg alloys are presented in Table II. This data illustrates the strengthening effect of increased Mg content under identical working conditions. The Cu addition results in a further strength increase, most probably the result of increased solid solution strengthening and more uniform substructure with the more uniform β distribution. As of this writing, transmission electron microscopy studies are being undertaken to examine the influence of such Cu and Mn additions

on β precipitation and substructure formation in these alloys. The Mn addition results in no further increase in strength, even though microstructural homogeneity is enhanced, and also results in some ductility loss.

B. THE INFLUENCE OF ANNEALING

The microstructural effect of the seven-day, 100°C anneal is presented in Figure 15. Annealing results in additional β precipitation in both the 8% Mg and 10% Mg alloys (Figs. 15a and b, respectively). The most notable effect in the 8% Mg alloy is increased precipitation on slip bands within grains, although there is also some increased grain boundary precipitation. The 10% Mg alloy is similarly affected, although increased grain boundary precipitation appears to be more the case with this material. In both alloys, the β particles have coarsened. The 10% Mg - 0.4% Cu alloy appears to be somewhat less affected by the anneal (Fig. 15c), although it now appears to pit in the electrolytic etch. The 10% Mg - 0.4% Cu - 0.5% Mn alloy (Fig. 15d) shows no discernable microstructural effect of this anneal.

The anneal resulted in reduced strength and increased ductility for the 8% Mg and 10% Mg alloys as illustrated in Figure 16. The effect is particularly notable for the 10% Mg alloy for which ductility increased by almost one-half as a result of the anneal. In contrast, the alloys containing the Cu addition or the Cu and Mn additions exhibited little

loss of strength and some decrease in ductility. The decreased strength and enhanced ductility noted for both the 8% Mg and 10% Mg alloys is consistent with the transmission electron microscopy results reported by Cadwell [Ref. 2]. A substructure not stabilized by the β particles, given their non-uniform distribution, would be able to recover and coarsen, resulting in reduced strength and improved ductility. The more uniform distribution of the β particles in the Cu or Cu and Mn containing alloys would result in a more stable substructure, less likely to coarsen during such an anneal. Figure 17 presents the same data as Figure 16, but in terms of the percentage of properties retained after the anneal.

The alloys may be ranked from those least affected to most affected by this annealing treatment. From the microstructure data such a ranking would be:

10% Mg - 0.4% Cu - 0.5% Mn; 10% Mg - 0.4% Cu; 8% Mg; 10% Mg.

Based on the mechanical property data as represented in Figure 17, the ranking is:

10% Mg - 0.4% Cu; 8% Mg; 10% Mg; 10% Mg - 0.4% Cu - 0.5% Mn.

The latter ranking presumes a material to be more adversely affected if ductility drops without gain in strength than if strength decreases with a corresponding increase in ductility. Hence these rankings are substantially different.

C. STRESS CORROSION TESTING

Susceptibility to stress corrosion cracking for the four warm rolled alloys and 7075-T6 was evaluated by alternate

immersion of stressed coupons in a 3.5% NaCl solution. Beam-type samples were loaded in three-point bending and additional samples were subject to a guided U-bend prior to testing. The alloys were all tested in the as-warm-rolled or as-received condition. Two additional tests were conducted on the warm rolled Al-Mg alloys after the annealing treatment. Prior to one test, the oxide layer developed during the anneal was removed. In the other test the oxide layer was left intact. The 7075-T6 was not annealed as the -T6 temper is the condition most prone to stress-corrosion cracking [Ref. 7]. Total test duration was 740 hours in all cases.

The results of these tests are summarized in Table III. The most significant observation is that the only failures attributable to stress corrosion occurred in three of the four 7075-T6 coupons subject to the guided U-bend. The 10% Mg - 0.4% Cu - 0.5% Mn alloy possessed insufficient ductility to conduct the guided U-bend test, either as-rolled or after the seven-day, 100°C anneal. The ductility loss incurred in the 10% Mg - 0.4% Cu alloy as a result of the anneal also precluded the guided U-bend test of this alloy in the annealed condition. From these data it can be concluded only that, for these test conditions, the warm-rolled Al-Mg alloys are not more susceptible to stress corrosion than the 7075-T6 alloy. In order to determine the effect of the stress corrosion test exposure, and to provide a ranking of the effect of this exposure, metallographic examination and mechanical testing of the test coupons was conducted at the conclusion of the 740 hour stress corrosion exposure.

1. Effect of Stress Corrosion Exposure on the As-Rolled and As-Received Condition

Longitudinal metallographic sections, from beam-type specimens of the five materials tested, are shown in Figure 18. In all cases, the sections include the exposed surface of the coupons in order to demonstrate the form of the stress corrosion attack on these materials. The 8% Mg alloy (Fig. 18a) was lightly affected, with a slight tendency for intergranular cracking extending from pitting. The 10% Mg alloy (Fig. 18b) exhibits a much more severe intergranular attack with extensive cracking spreading from surface pitting in the manner of exfoliation. This form of attack also occurs for the 10% Mg - 0.4% Cu alloy, although the tips of the intergranular corrosion cracks appear blunted in this alloy, as shown in Figure 18c. The last of the warm-worked alloys examined, the 10% Mg - 0.4% Cu - 0.5% Mg material, was subject to an entirely different form of attack. Figure 18d shows shallow, rounded pits to have formed, and no tendency for intergranular cracks to spread from these pits. The 7075-T6 alloy (Fig. 18e) was severely affected by the stress corrosion exposure, with intergranular cracking spreading from irregular pitting in the alloy.

The results of mechanical testing following this stress corrosion test are summarized in Figure 19. The strength and ductility of the 8% Mg and 10% Mg alloys were little affected by the exposure, the 8% Mg alloy being the least affected of

all. The tensile properties, especially ductility, of the remaining alloys were more severely reduced by the exposure. To aid in ranking the effect of the exposure and hence the severity of the attack, these data were replotted in Figure 20 in terms of percentage of property (strength and ductility) retained after exposure as compared to beforehand.

As before, the alloys may be ranked first on the basis of the apparent severity of the microstructural degradation from least to most severely degraded:

8% Mg; 10% Mg - 0.4% Cu -0.5% Mn; 10% Mg; 10% Mg - 0.4% Cu; 7075-T6

Similarly, the ranking based on mechanical property degradation, from least to most severely affected, is:

8% Mg; 10% Mg; 10% Mg - 0.4% Cu - 0.5% Mn; 10% Mg - 0.4% Cu; 7075-T6

These rankings differ only in the reversal of the positions of the 10% Mg alloy and the 10% Mg - 0.4% Cu - 0.5% Mn alloy.

The rankings given above are consistent with the result of the stress-corrosion test itself, in that the 7075-T6 alloy, most severely degraded by the stress corrosion exposure, was also the only alloy for which any actual stress corrosion failure occurred. The warm rolled alloys which exfoliated during exposure were subsequently pulled in tension parallel to the original rolling direction, which is the long direction of the grains. For this reason, the severity of the cracking mode noted for the 10% Mg and 10% Mg - 0.4% Cu alloys is not reflected

in these rankings. Such cracking would certainly result in a more pronounced effect on properties if measured in the through-thickness direction. In contrast, the Mn containing alloy, which exhibited only rounded pitting, would not likely be as strongly affected in the through thickness orientation.

As noted previously, the Mn-containing alloy was processed differently in that an additional, solution treatment was necessary. The microstructural homogeneity evident in this alloy may result in part from this as well as the Mn addition itself. In turn, the altered form of environmental attack may also be the result of the different process as well as the Mn addition. This suggests study of the influence of solution treatment temperature for the other alloys examined here to determine the separate effects of this parameter and the Mn addition. As noted by Johnson [Ref. 1], this alloy may also be processed to considerably higher tensile strength (up to 100 KSI (690 MPa)) by reduced warm rolling temperatures. The influence of this aspect of this processing method also was not investigated here.

2. Effect of the Anneal on the Response to the Stress-Corrosion Exposure

This second series of tests was conducted in two parts. As noted, half of the annealed test coupons were polished prior to exposure while half were tested with the oxidation resulting from the anneal left intact. Examination of metallographic samples from both parts of this test revealed no difference

in the form of the attack resulting from the presence or absence of the oxide. Figure 21 shows micrographs of longitudinal sections like those discussed previously in Figure 18. This series is from the test of the annealed and polished series. Comparison of these figures reveals a similar form of degradation for three of the four alloys. The 8% Mg alloy is still relatively unaffected by exposure (Fig. 21a). The 10% Mg alloy (Fig. 21b) again exhibits intergranular cracking and exfoliation although the cracking is somewhat blunted in comparison to the as-rolled condition. This micrograph also shows a section through a blister and the exfoliation occurring underneath the blister. The 10% Mg - 0.4% Cu alloy (Fig. 21c) also exhibits intergranular cracking as before.

In the as-rolled condition, the 10% Mg - 0.4% Cu - 0.5% Mn alloy was subject to formation of shallow rounded pits but with no cracking extending from these pits. The anneal before stress corrosion exposure results in better surface retention during exposure with almost no pitting in evidence and, again, no exfoliation (Fig. 21d). This result is consistent with the apparent stability of the microstructure of this material during the anneal.

The absence or presence of the oxide layer during stress corrosion testing did affect the resultant mechanical properties, most notably those of the 10% Mg alloy. Figure 22 compares the annealed mechanical properties to those of the annealed and exposed test coupons. The oxide layer formed in

the anneal had been removed prior to exposure for this sample series. The 10% Mg alloy now has been severely degraded in ductility. As noted microstructurally, this alloy also showed evidence of blistering in addition to exfoliation and this additional factor is likely involved in the large degradation of ductility observed. The 8% Mg alloy exhibits little degradation of properties in contrast to the higher Mg alloy. The 10% Mg - 0.4% Cu and the 10% Mg - 0.4% Cu - 0.5% Mn alloys are affected to a slightly lesser extent by exposure after annealing than by exposure in the as-rolled condition. This is better seen by comparing Figure 23 to Figure 20. Both of these figures represent data as percentage of property retained after exposure, Figure 23 for the annealed (and polished) condition and Figure 20 for the as-received or as-rolled condition. This comparison also illustrates the severe degradation experienced by the 10% Mg alloy as a result of annealing.

Mechanical test results for the test series annealed and exposed with the oxide layer intact are given in Figure 24. Examination of these data reveal slightly better retention of mechanical properties with the oxide layer intact when compared to the data of Figure 22. The same conclusion is reached if data is represented in terms of percentage of property retained after exposure, as shown in Figure 25, and if comparison is made to the data of Figure 23.

These results again suggest a ranking, now considering only the four warm rolled alloys, of the effect of stress corrosion

exposure on the annealed condition. Based on the microstructural data a ranking from lease affected to most affected is:

8% Mg; 10% Mg - 0.4% Cu - 0.5% Mn; 10% Mg - 0.4% Cu; 10% Mg.

Ranking similarly based on the mechanical test results of Figure 23, for the annealed and polished test series is:

8% Mg; 10% Mg; 10% Mg - 0.4% Cu; 10% Mg - 0.4% Cu - 0.5% Mn; 10% Mg.

Only the position of the 10% Mg alloy is changed in such a ranking if the annealed, unpolished series (Fig. 25) is considered. This alloy would now fall after the 8% Mg alloy:

8% Mg; 10% Mg; 10% Mg - 0.4% Cu; 10% Mg - 0.4% Cu - 0.5% Mn

As noted previously, the 7075-T6 alloy was not annealed and therefore not tested in the annealed condition. The anneal was intended to sensitize the warm-rolled alloys and the -T6 temper is already the most sensitive to stress corrosion for the 7075 alloy [Ref. 7]. Comparison of the response of the annealed materials can therefore be made to the data for the 7075-T6 previously given in Figures 19 and 20. Such comparison immediately reveals that the warm-rolled alloys are still not as severely degraded by exposure of the materials after the sensitizing anneal as the 7075-T6 alloy, exposed in the as-received condition.

3. Summary of the Effect of Exposure: Consideration of the Alloys

Figures 26-30 summarize the results of the evaluation of these alloys following the stress corrosion test. These data were examined to provide a final, overall ranking of the effect of stress corrosion exposure. These data are presented in order of least affected to most affected as:

8% Mg (Fig. 26); 10% Mg - 0.4% Cu - 0.5% Mn (Fig. 27);
10% Mg - 0.4% Cu (Fig. 28); 10% Mg (Fig. 29); 7075-T6 (Fig. 30)

The position of the 8% Mg alloy (Fig. 27) in these rankings has remained the same throughout this study. This alloy, in general, is little affected by the stress corrosion exposure with tensile strength unaffected by exposure and with loss of small fraction, typically less than one-tenth, of the ductility possessed prior to exposure. It has been noted, however, that the grain structure of this alloy is highly elongated and that some precipitation has occurred in the grain boundaries. However, this precipitation is discontinuous and it is possible that the extensive warm rolling has resulted in lesser local concentration gradients in material nearby grain boundaries than would be the case if such precipitation were the result of diffusional processes alone. In this sense, such extensive warm working as utilized with these materials may provide less stress corrosion sensitive microstructures than might be expected in such high Mg alloys by virtue of microstructural homogenization.

The data for the 10% Mg - 0.4% Cu - 0.5% Mn alloy, summarized in Figure 27, also reveals this materail as not severely degraded by stress corrosion exposure. This alloy is the highest strength, lowest ductility alloy of the series of warm-rolled alloys evaluated and, as such, might have been expected to show more severe degradation than actually observed. Two features distinguish this material. One is the alloying additions, Cu and Mn in particular, and the other is the modified solution treatment utilized. The combination of these factors has resulted in a very homogeneous microstructure after warm rolling. This observation suggests closer attention both to the solution treatment prior to warm rolling and to the effects of the Mn addition.

The 10% Mg - 0.4% Cu alloy (Fig. 28) exhibits an anomalous effect. In the as-rolled condition, the ductility of this material was reduced from 7.5% elongation to about 2% elongation as a result of the exposure. Even though annealing reduced ductility slightly, to about 6.5% elongation, a higher ductility (about 3.5%) was noted after exposure than after exposure as-rolled. This suggests that the anneal had a desensitizing, rather than a sensitizing, effect in this case. This is only a very tentative observation at this point given the relatively low ductilities noted after exposure. This again, points out the possibilities of this warm-rolling method for development of stress corrosion cracking resistant microstructure in these high Mg, Al-Mg alloys, in conjunction with relatively high strength.

The final warm-rolled alloy in this ranking is the 10% Mg material (Fig. 29). This evaluation is dictated by the severe degradation of properties noted by exposure after annealing and as well the tendency toward intergranular cracking and exfoliation. This alloy possesses consistently higher ductility than others evaluated in this study. However, the method of post-exposure testing does not reflect the severity of degradation observed in that the coupons are stressed parallel to the grain orientation and therefore the cracks developed are also stressed parallel to rather than perpendicular to the crack plane. A similar observation may be made regarding the 7075-T6 alloy (Fig. 30), however, in that post-exposure testing also was parallel to the crack plane of the intergranular cracks observed. This alloy was generally the most severely affected in any case, especially with regard to loss of ductility after exposure.

D. THE GENERAL CORROSION TEST

The effects of general corrosion exposure in a salt-spray environment were evaluated by periodic weight loss measurement during the exposure. Two separate, nominally identical, tests were conducted. The results of these tests are presented in Figures 31 and 32. These data show identical trends in behavior for these two tests; however, weight losses are larger in Run 2, suggesting some possible differences in the actual spray environment.

The 8% Mg and 10% Mg exhibited almost no weight loss over the 1000 hour duration of the test. The alloys with the Cu and Cu-Mn additions behave in a similar manner with the onset of attack at about 100 hours and an acceleration in weight loss thereafter. These latter alloys are intermediate in behavior to the binary (8 or 10% Mg) alloys and the 7075-T6, which consistently exhibits the greatest weight loss.

An additional test of these alloys consisted of exposure to an actual marine atmosphere by exposing a series of test coupons aboard the R/V ACANIA, a research vessel with home port in Monterey, California. This test was run for 1500 hours, at which point weight loss measurements were made. These data are presented in Table IV along with the tabular data for the accelerated salt-spray test. These alloys show an identical trend in weight loss to that observed in the salt-spray test are equivalent to weight loss in the accelerated test at times between 50 and 100 hours.

The general appearance of a series of the salt-spray test coupons is shown in the macrophotographs of Figure 33. These were obtained using oblique lighting; hence, unaffected regions of the originally-polished coupons appear dark. The 8% Mg and 10% Mg alloys exhibit the least corrosive attack, although the 10 Mg alloy is somewhat more degraded than the 8% Mg alloy. The 10% Mg - 0.4% Cu and the 10% Mg - 0.4 Cu - 0.5% Mn alloys both show more extensive surface degradation and the 7075-T6 alloy shows the most extensive degradation with numerous, deep pits.

Figure 34 shows a series of longitudinal metallographic sections including the ends as well as surfaces of these salf-spray test coupons. Again, the 8% Mg alloy (Fig. 34a) is only lightly attacked. However, the 10% Mg alloy (Fig. 34b) is seen to experience intergranular attack from the end near the surface, leading to exfoliation. Close examination also reveals numerous intergranular cracks penetrating from the exposed end of the sample and some cracking from blisters. The 10% Mg - 0.4% Cu alloy exhibits no intergranular attack (as shown in Fig. 34c) whereas under stress-corrosion exposure such attack was observed. Further, no pitting is observed; given that weight loss has occurred, the corrosion attack is taking place uniformly for this alloy. Figure 34d illustrates intergranular attack from the exposed ends of the test coupon for the 10% Mg - 0.4% Cu - 0.5% Mn alloy, however, the cracking is characterized by very blunt crack tips. The 7075-T6 alloy is seen to exhibit extensive intergranular attack from the exposed ends of the test coupon as shown in Figure 34d, and extensive pitting with some intergranular cracking spreading from this pitting.

The above observations may be used, as before, to rank these alloys, now with regard to their resistance to general corrosion. Based on weight loss measurements, the ranking, from most resistant to least is:

8% Mg and 10% Mg; 10% Mg - 0.4% Cu and 10% Mg - 0.4% Cu - 0.5% Mn; 7075-T6

Since metallographic study indicates intergranular attack in two cases, these rankings would necessarily change to:

8% Mg; 10% Mg - 0.4% Cu; 10% Mg; 10% Mg - 0.4% Cu - 0.5% Mn; 7075-T6

The alteration in order reflects the observation of blistering and intergranular attack in the 10% Mg alloy, even given the small weight loss noted for this alloy.

The addition of Cu to Al-Mg alloys is done to enhance general corrosion resistance and retard pitting [Ref. 18]. In these alloys, however, the binary 8% Mg and 10% Mg exhibit the greatest resistance to weight loss and pitting in the salt-spray test. It should be noted that Cu additions to conventional Al-Mg alloys are typically less than 0.2% while the Cu addition in both alloys of this study is about 0.4%. Similarly, Tomashov [Ref. 19] notes that Mn additions up to 2.0% to Al alloys result in highly corrosion and stress-corrosion resistant alloys. In this research, 0.5% Mn addition to the 10% Mg - 0.4% Cu composition resulted in a material less resistant to stress-corrosion degradation than a binary 8% Mg alloy, although considerably higher in strength. This alloy similarly is less corrosion resistant than the binary 8% Mg and 10% Mg alloys.

IV. CONCLUSIONS AND RECOMMENDATIONS

A. CONCLUSIONS

1. The alloys evaluated in this test are less susceptible to general and stress corrosion than the control alloy 7075-T6.

2. The Thermomechanically processed 8% Mg alloy exhibits the best overall resistance to general and stress corrosion.

3. The Thermomechanically processed 10% Mg alloy exhibits blistering and exfoliation in both general and stress corrosion environments, and it is severely degraded by the sensitizing anneal. This is the result of insufficient homogenization of the β phase.

4. The addition of Cu has a homogenizing and stabilizing effect on the Thermomechanically processed 10% Mg alloy.

5. The addition of Cu and Mn in conjunction with the additional solution treatment at 490°C for 3 hours results in the most homogeneous and stable microstructure in the Thermomechanically processed 10% Mg alloys.

B. RECOMMENDATIONS

1. Further study of processing and alloying variables, especially the solution treatment and Mn addition, should be conducted. In the solution treatment stage, effects of increased time and temperature should be considered. Final rolling temperature as well should be included as it has a strong effect on resultant strength.

2. Extended testing to failure should be conducted to validate these results. This should be in conjunction with environmental exposure of these materials.

TABLE I.

Alloy Composition

ALUMINIUM ALLOY DESIGNATION	Mg	Cu	Mn	Cr	Zn	Si	Fe	Ti	Be
7075-T6	2.5	1.6	--	0.23	5.6	--	--	--	--
501302 A	10.4	0.43	0.52	--	--	--	0.03	0.01	0.0002
501301 A	10.3	0.41	--	--	--	0.01	0.03	0.01	0.0002
501299 A	10.2	--	--	--	--	0.01	0.03	0.01	0.0002
486250 A	8.0	--	--	--	--	--	--	--	--

TABLE II.

Mechanical Properties of the As-Received 7075-T6 and the Warm Rolled Alloys

ALLOY	YIELD STRENGTH KSI (MPa)	TENSILE STRENGTH KSI (MPa)	DUCTILITY, PCT ELONGATION
7075-T6	83 (573)	89 (614)	11
10% Mg - 0.4% Cu - 0.5% Mn	66 (455)	80 (552)	5.6
10% Mg - 0.4% Cu	65 (449)	80 (552)	7.6
10% Mg	59 (407)	73 (504)	10.8
8% Mg	50 (345)	61 (421)	9.5

TABLE III.

Summary of Stress Corrosion Results in 3.5% NaCl Solution, Alternate Immersion for 740 Hours

CONDITION BEFORE EXPOSURE	SAMPLE GEOMETRY	7075-T6	10% Mg .43% Cu .52% Mn	10% Mg .43% Cu	10% Mg	8% Mg
as-rolled/ as-received	U bend	F F F -	* *	- -	- -	- -
	bent beam	- - - - - - - - - - - -	- - - - - -	- - - - - -	- - - - - -	- - - - - -
annealed polished	U bend		@	@	-	-
	bent beam		- - -	- - -	- - -	- - -
annealed unpolished	U bend		@	@	-	-
	bent beam		- - -	- - -	- - -	- - -

'F' indicates failure during the tes. '*' indicates that this alloy had insufficient ductility to form a U bend. '@' indicates that these alloys had insufficient ductility to form a U bend after the anneal. '-' indicates no failure during the 740 hour exposure.

TABLE IV.

Weight Loss Data Measured After Exposure Times As Noted

EXPOSURE TIME (hours)	RUN NO.	WEIGHT LOST (mg) 7075-T6	10% Mg, Cu, Mn	10% Mg, Cu	10% Mg	8% Mg
5	1	1.5	-.1	.5	.5	-.8
	2	2.3	-.3	1.1	2.1	-.2
10	1	1.2	.4	1.4	1.6	-.8
	2	3.3	.8	1.5	2.0	-.3
50	1	7.7	.6	2.8	1.4	-.4
	2	14.0	1.1	3.0	1.9	1.2
100	1	13.3	2.8	3.5	2.2	1.4
	2	29.7	5.2	7.3	2.7	1.7
500	1	40.3	10.2	21.8	2.7	1.8
	2	75.4	31.7	23.7	3.3	3.3
1000	1	65.7	30.1	27.0	3.2	3.8
	2	110.0	54.0	74.7	5.8	5.6
1500 hour exposure aboard the R/V ACANIA		8.86	4.93	4.36	1.26	1.18

Samples were cleaned with 70% nitric acid and air dryed before weighing.
Note: The actual exposure data falls in at the 50 to 100 hour point of the data for the 3.5% NaCl spray test.

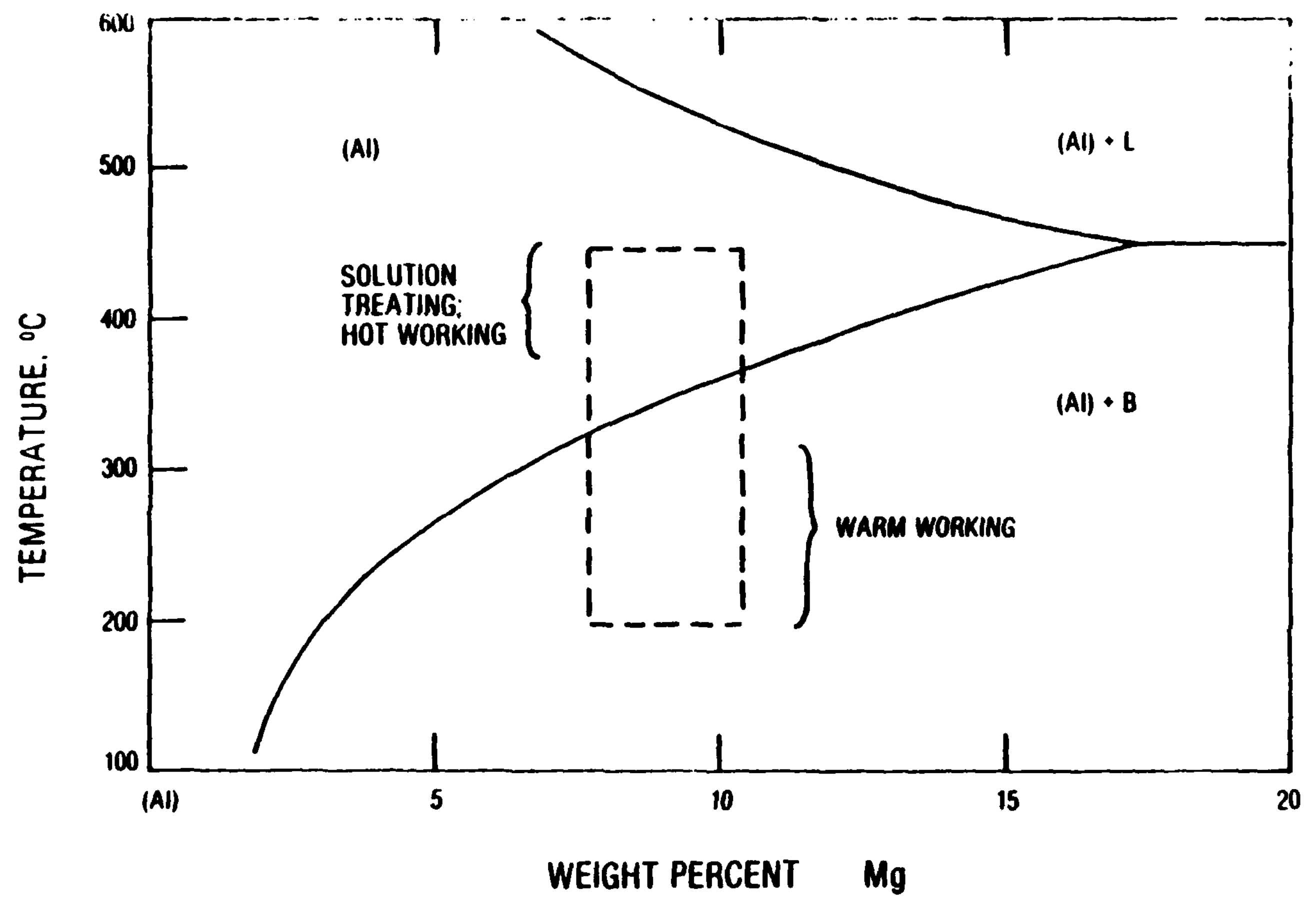


Figure 1. A partial Aluminum-Magnesium binary phase diagram, applicable to the Aluminum-Magnesium alloys of this work, illustrating the hot and warm working temperature ranges for eight to ten weight percent Magnesium alloys.

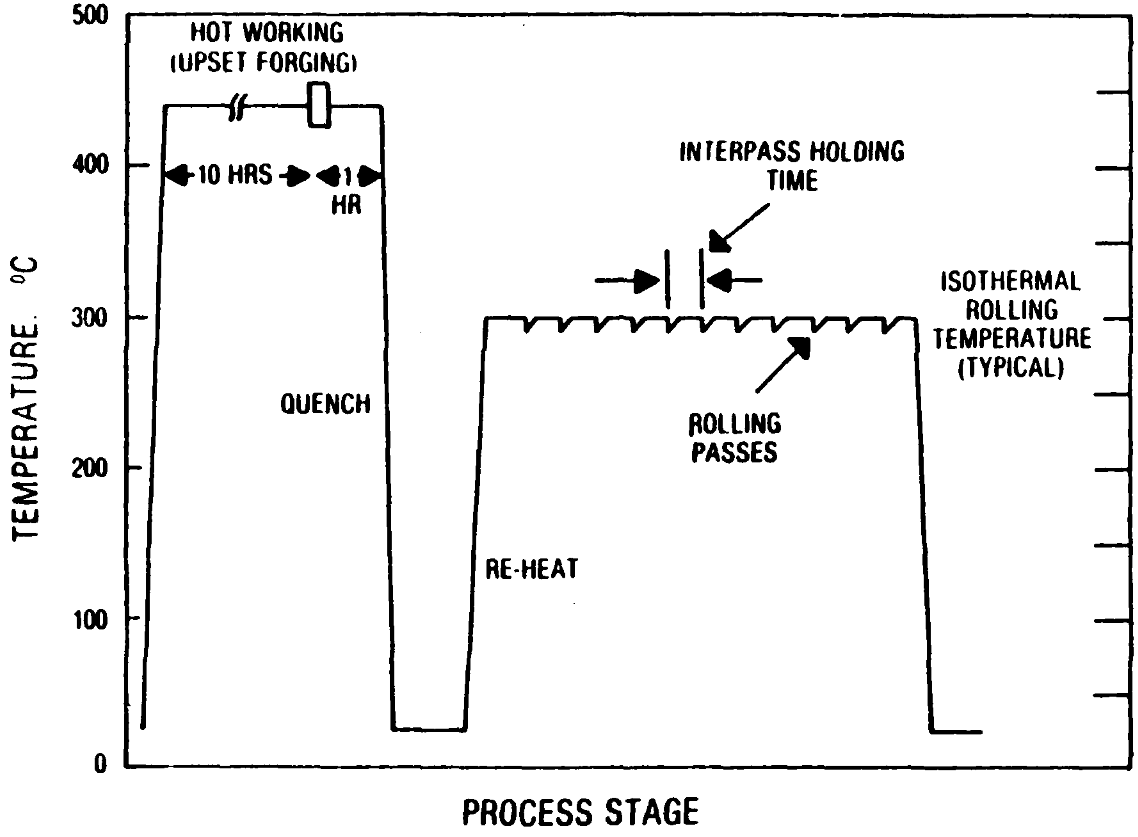


Figure 2. A schematic representation of the processing method applied to the high-Magnesium, Aluminum-Magnesium alloys of this work.

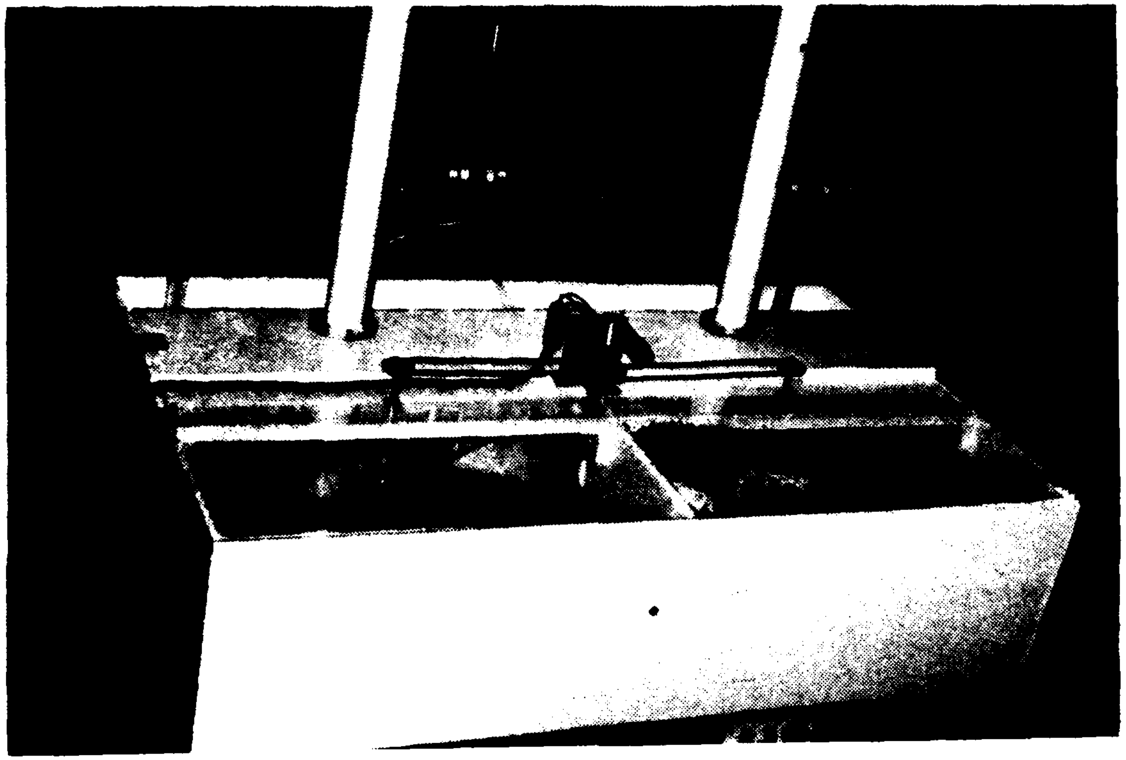


Figure 3. General and stress corrosion test tank. In the forground are the two general corrosion spray chambers. In the background are two stress corrosion alternate immersion racks, and in the center are the exhaust ducts.

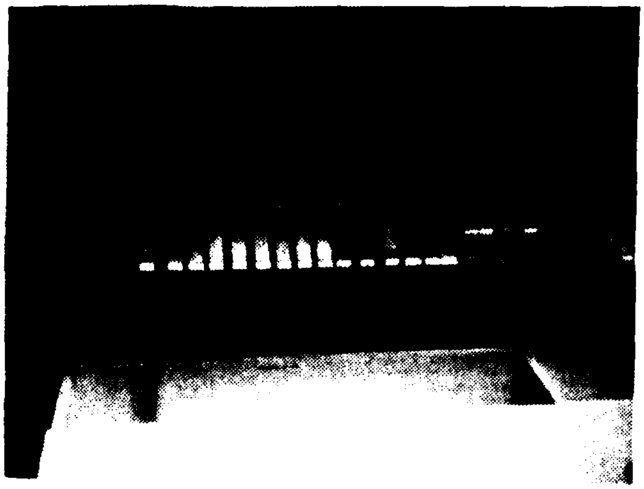


Figure 4. Stress corrosion alternate immersion rack.

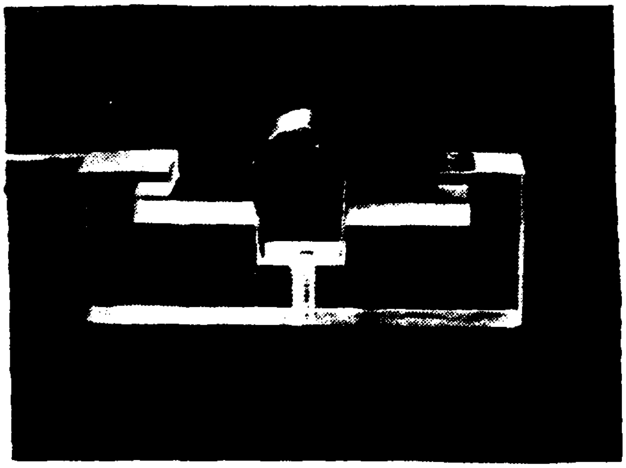


Figure 5. U-bend stress corrosion sample and holder.

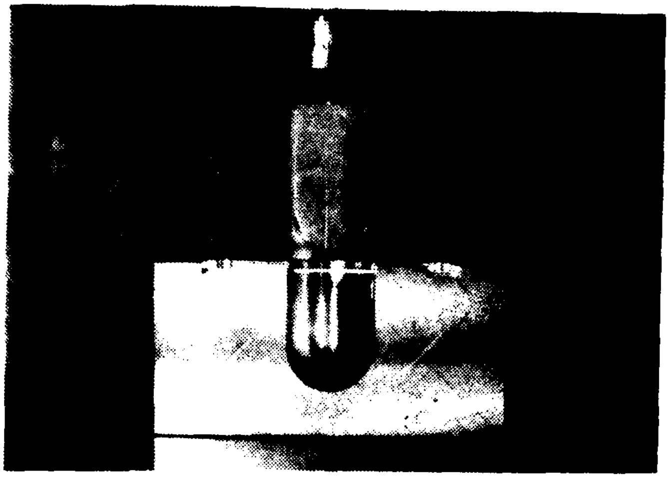


Figure 6. Apparatus for making the guided U-bend samples.

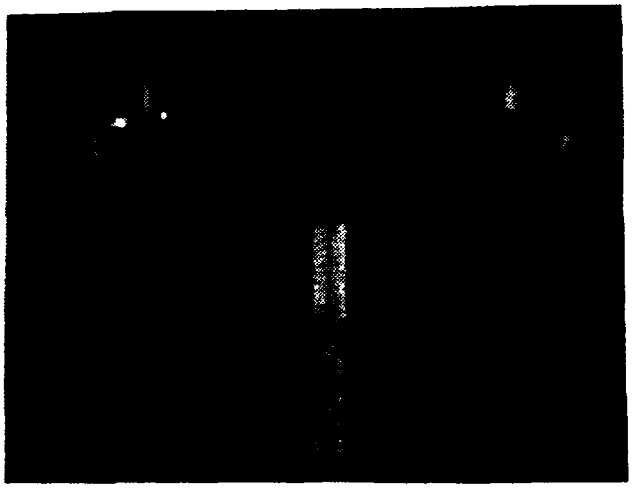


Figure 7. Three-point bent-beam stress corrosion sample and holder.

Figure 8. Apparatus for measuring the surface deflection on the three-point bent-beam samples.

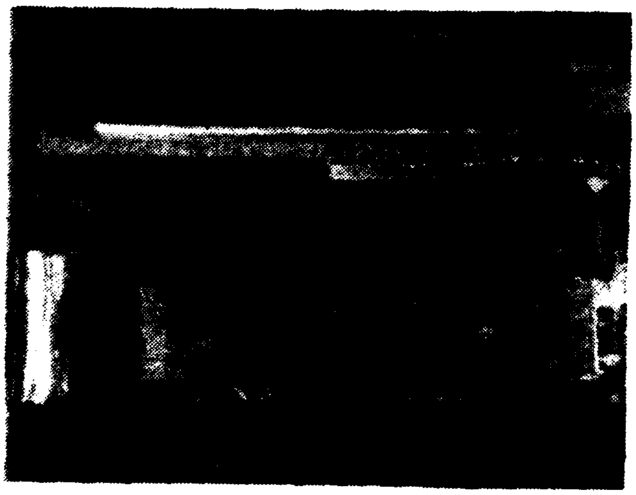


Figure 9. General corrosion sample and holder.

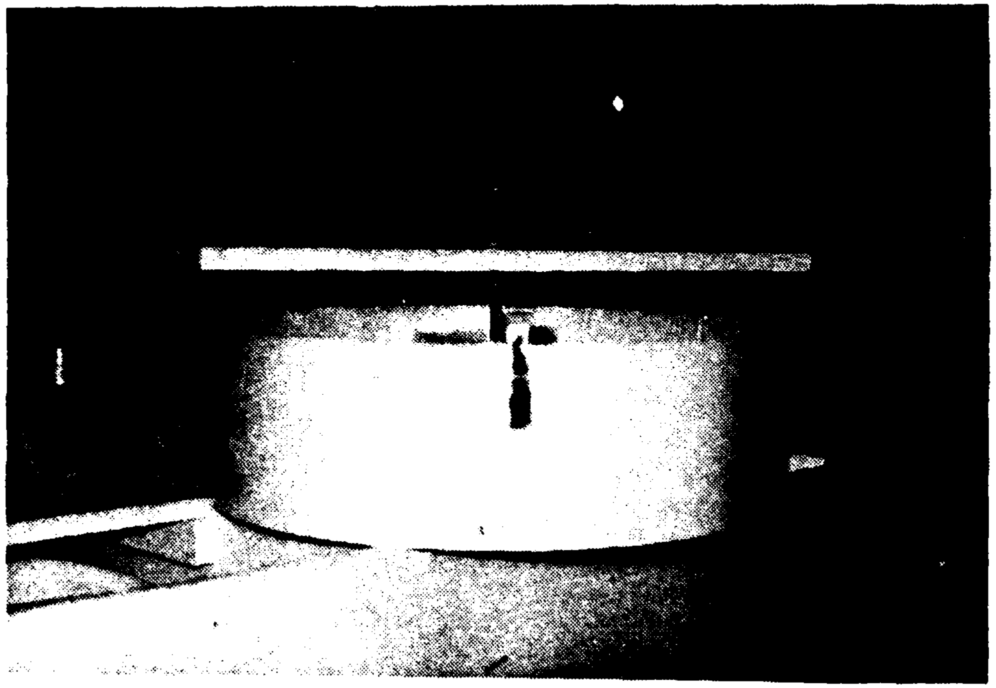


Figure 10. One rim assembly with a general corrosion sample and holder in place.

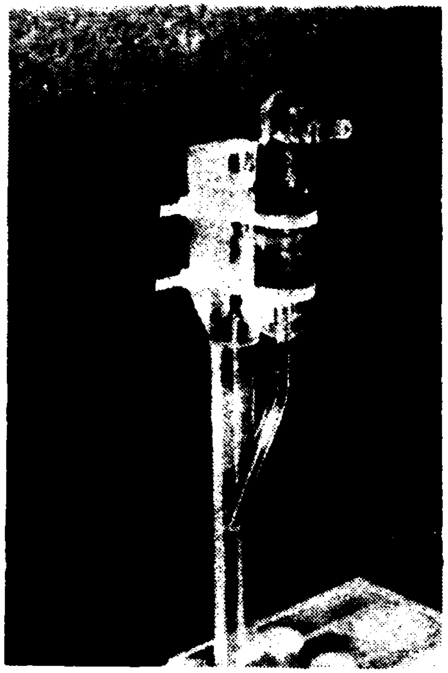


Figure 11. Spray nozzle.

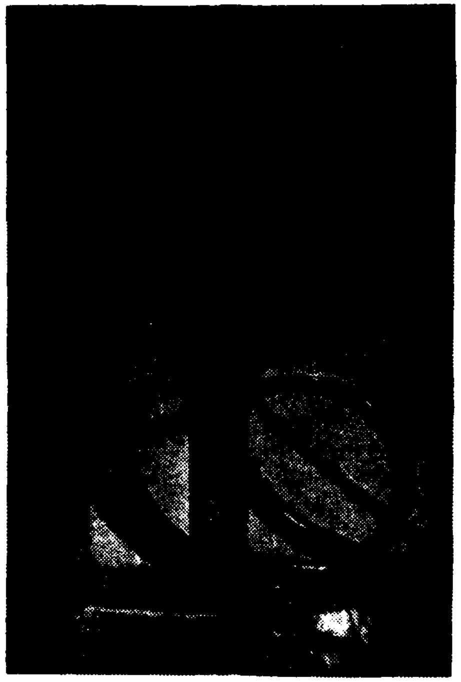


Figure 12. Typical installation with hoses for salt solution and air.

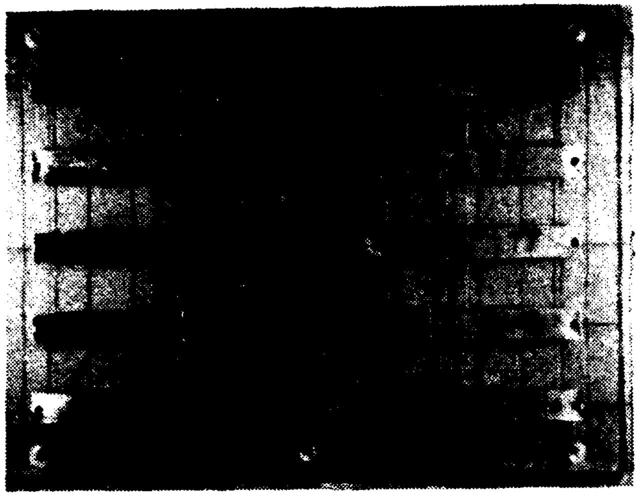
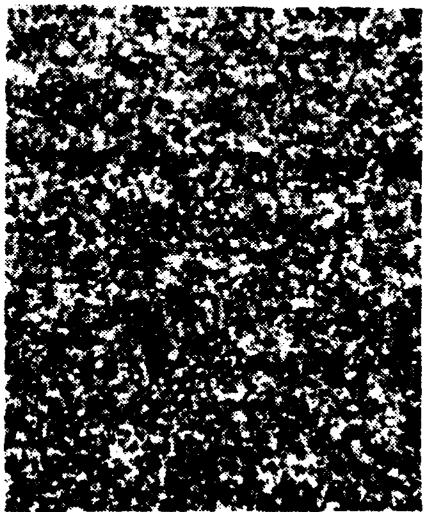


Figure 13. Exposure rack and samples (three of each alloy) used aboard the R/V Acania.

(a)

(b)

(c)



(d)

Figure 14. The as rolled material: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn. Etched at 20 volts for 20 seconds in Barkers reagent. 500X.

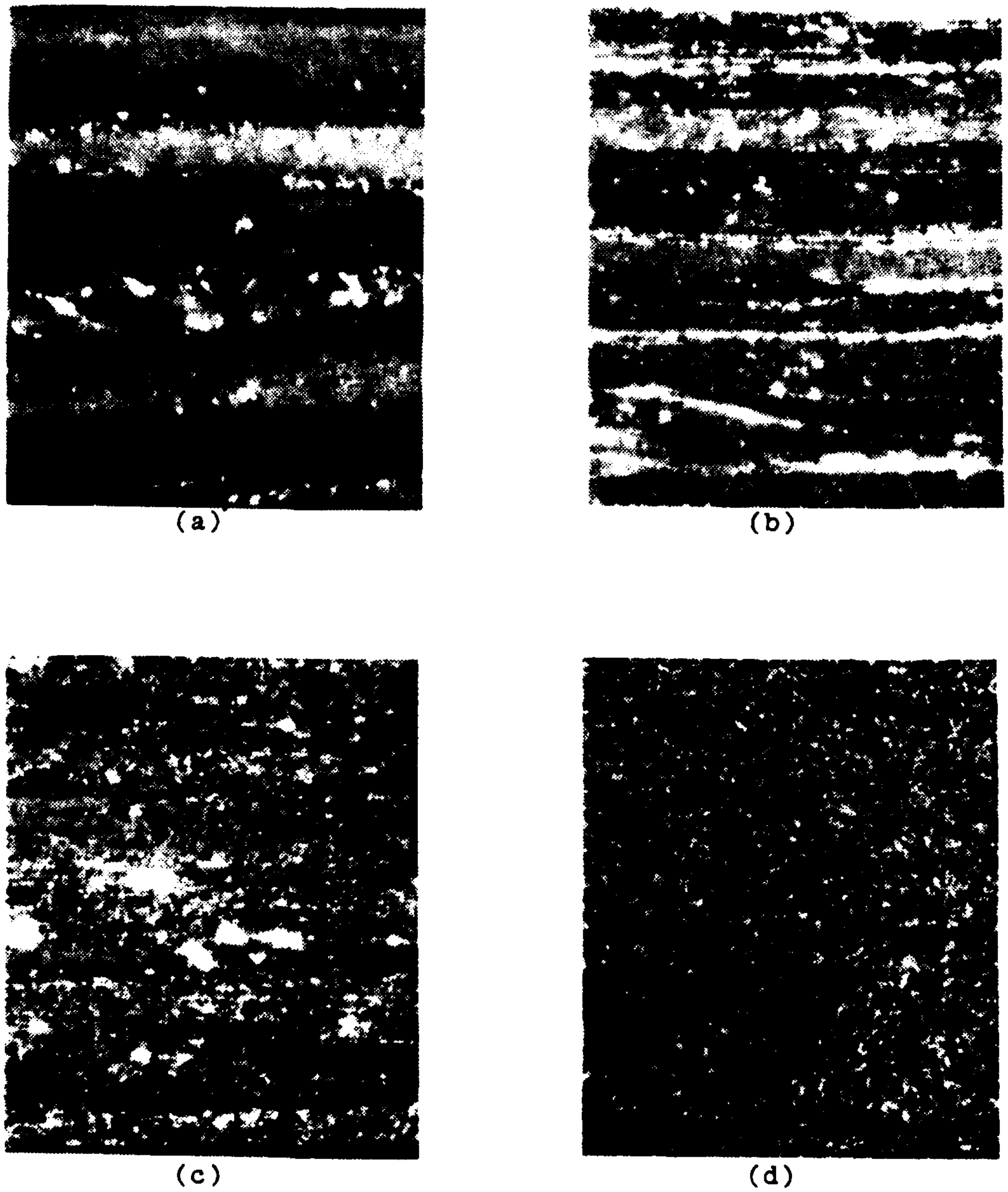


Figure 15. The as rolled material after a seven day 100 $^{\circ}C$ anneal. a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu d) 10% Mg, 0.4% Cu, 0.5% Mn. Etched at 20 volts for 20 seconds in Barkers Reagent. 500X.

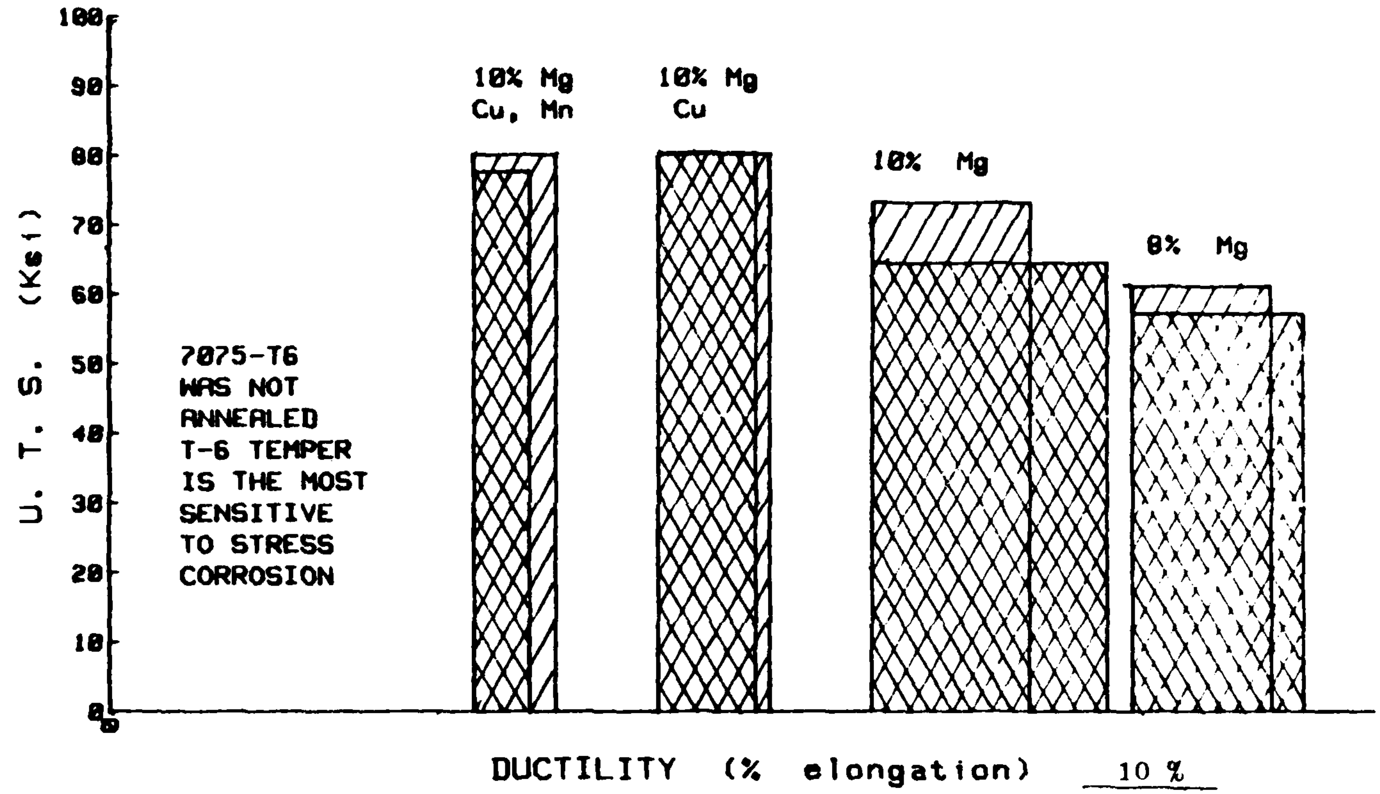


Figure 16. Comparison of the mechanical properties of the Al-Mg alloys in the as rolled condition to the properties after a seven-day 100 °C anneal.

AS ROLLED ANNEALED

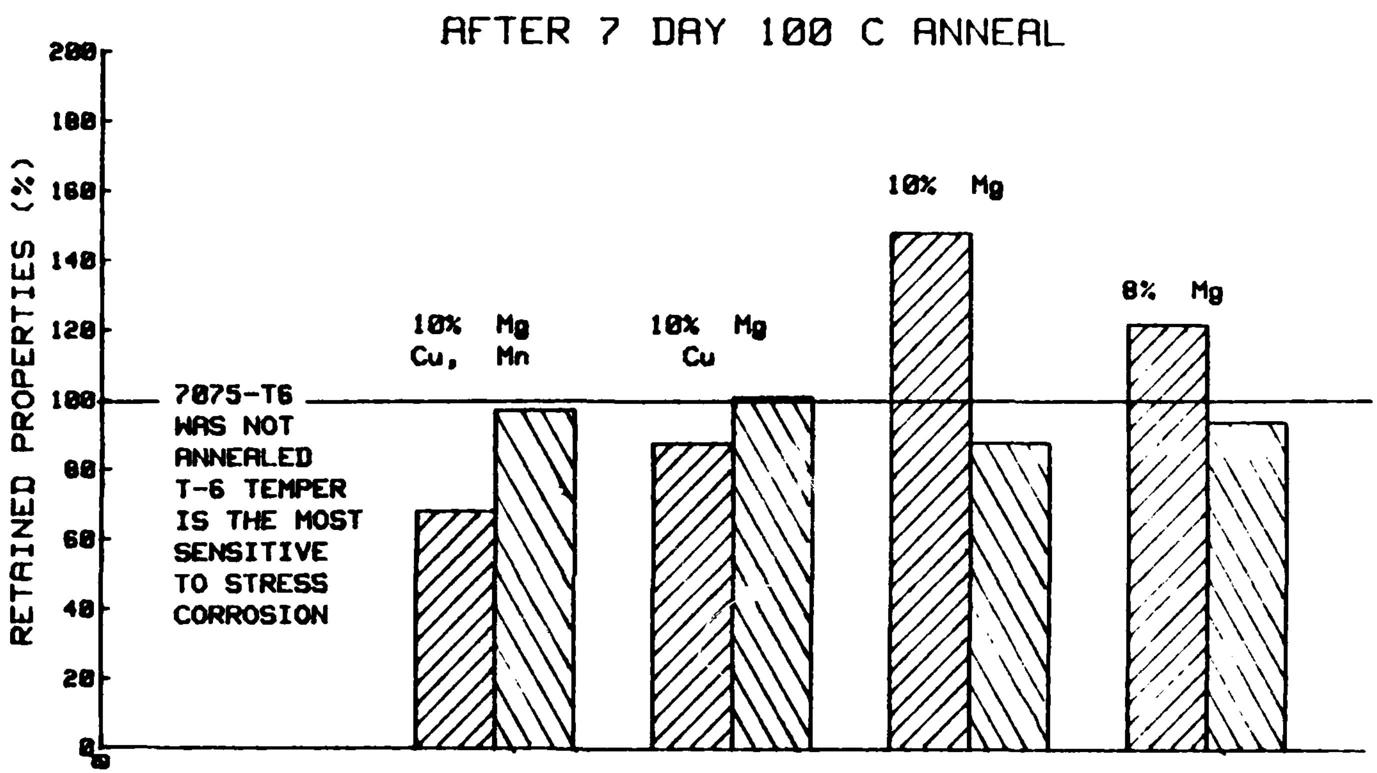
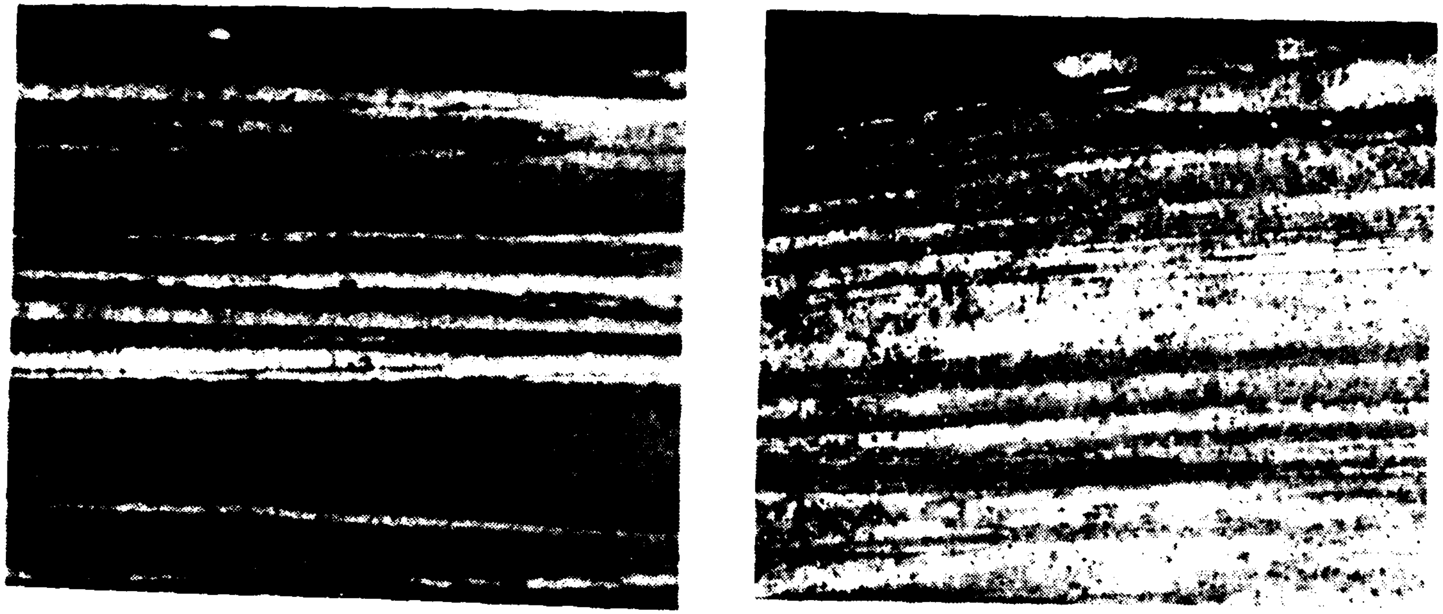


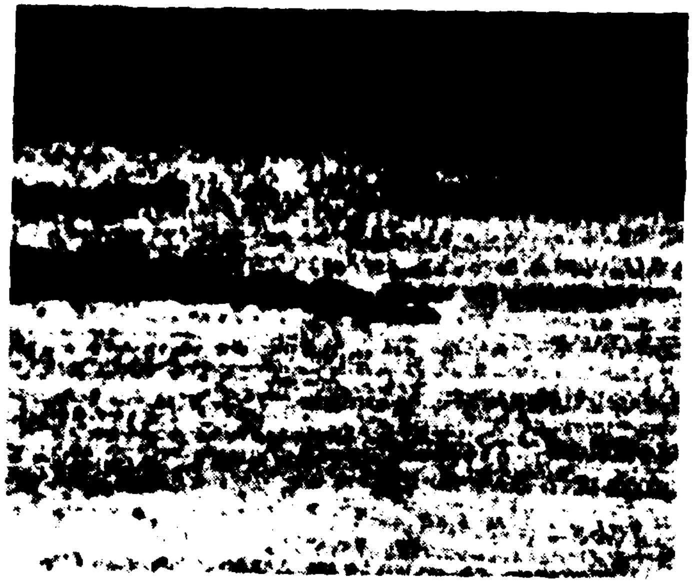
Figure 17. Effect of a seven-day 100 ^{O}C anneal on the as-rolled Al-Mg alloys. The data is represented as percentage of property retained, refered to the as rolled condition.

DUCTILITY ULTIMATE TENSILE STRENGTH



(a) (b)

Figure 18. The as rolled/as received material after 740 hours of alternate immersion stress corrosion exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched for 20 seconds at 20 volts in Barkers reagent, the 7075-T6 was etched in Kellers etch for 20 seconds.



(c)

(d)

Figure 18. (continued) The as rolled/as received material after 740 hours of alternate immersion stress corrosion exposure: a)8% Mg, b) 10%Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched for 20 seconds at 20 volts in Barkers reagent, 7075-T6 was etched for 20 seconds in Kellers etch. 500X.

(e)

Figure 18. (continued) The as rolled/as received material after 740 hours of aternate immersion stress corrosion exposure: a) 8% Mg, b) 10% Mg c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched for 20 seconds at 20 volts in Barkers reagent, 7075-T6 was etched in Kellers etch for 20 seconds. 500X.

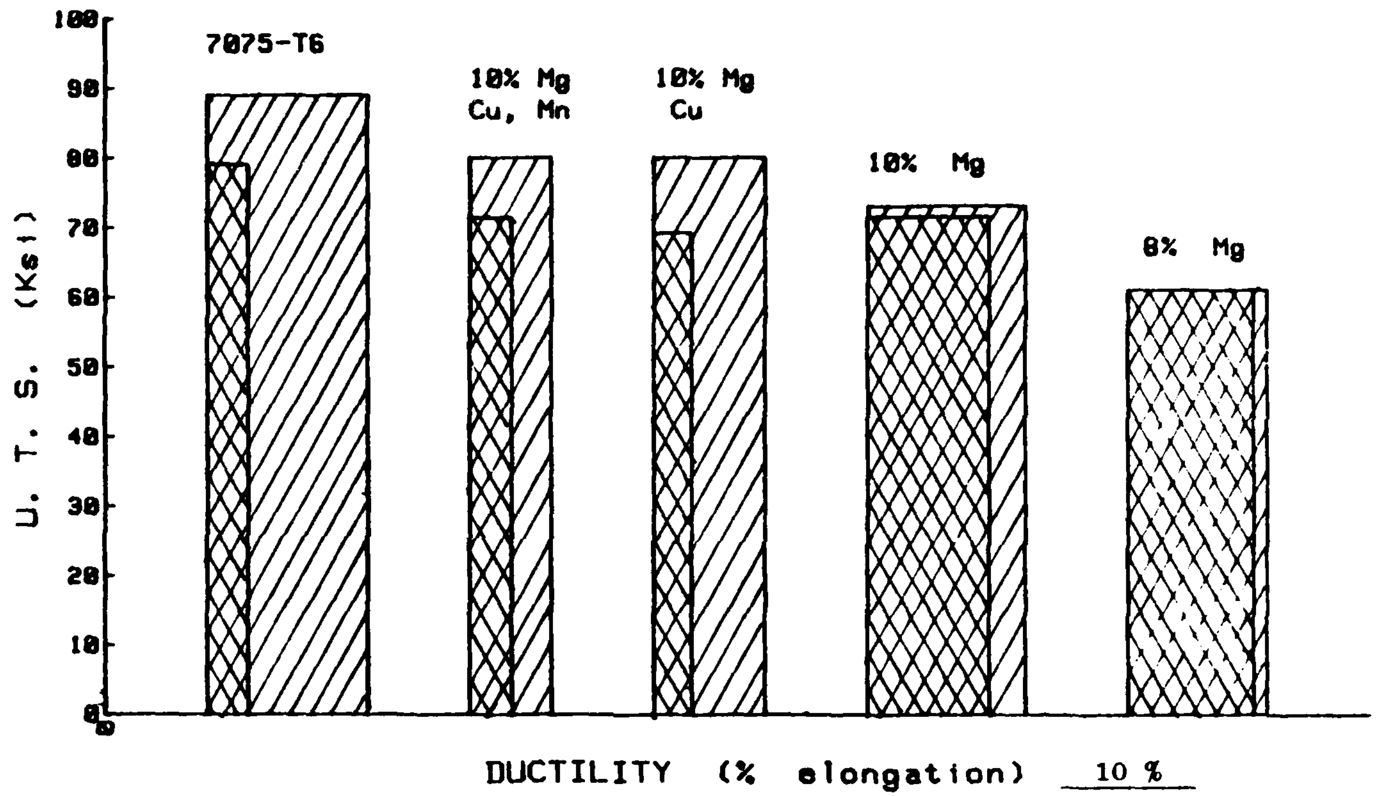


Figure 19. Influence of stress corrosion exposure on the mechanical properties of the as-received 7075-T6 and the as rolled Al-Mg alloys. The test samples were polished prior to exposure.

AS-RECEIVED/AS-ROLLED ▨ STRESS CORRODED ▩

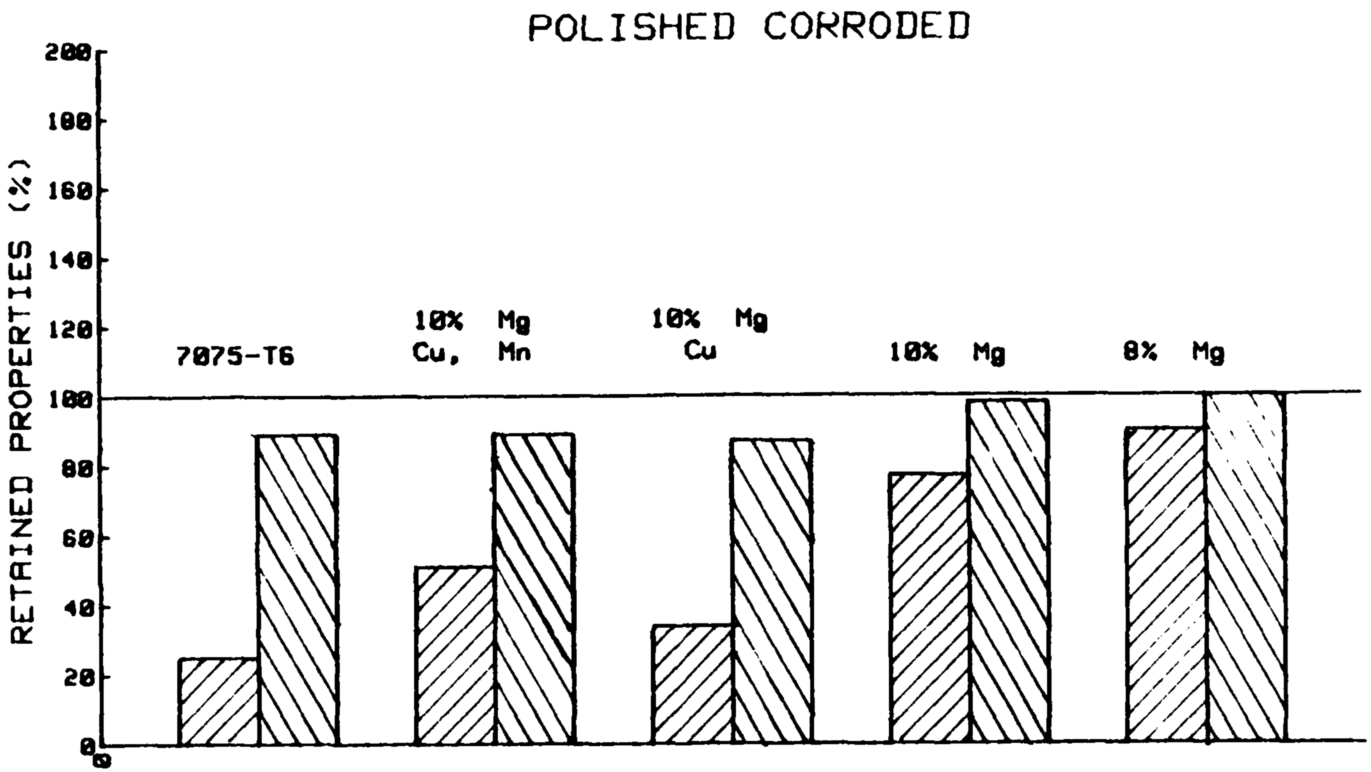
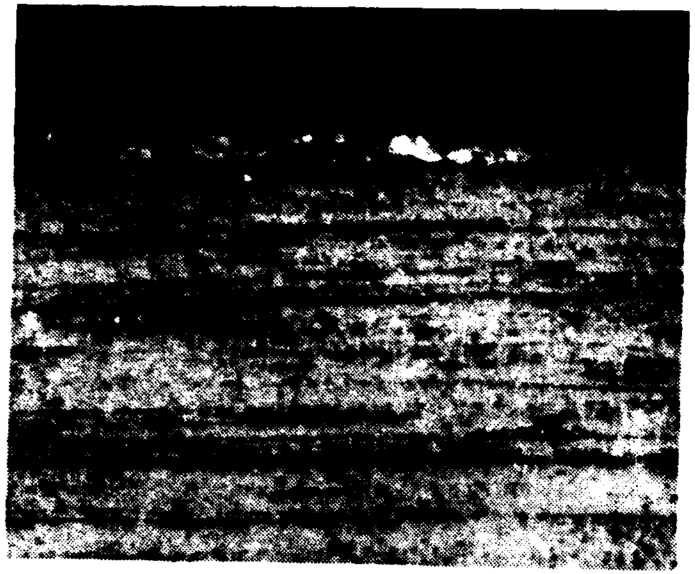


Figure 20. Effects of stress corrosion exposure on the as-received 7075-T6, and the as-rolled condition for the Al-Mg alloys. The data is represented as percentage of property retained, refered to the as-received or as-rolled condition. The test samples were polished prior to exposure.

DUCTILITY ULTIMATE TENSILE STRENGTH



(a)

(b)

Figure 21. The annnealed Al-Mg alloys after 740 hours of alternate immersion stress corrosion exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu d) 10% Mg, 0.4% Cu, 0.5% Mn. Etched at 20 volts for 20 seconds in Barkers reagent. 500X.

(c)

(d)

Figure 21. (continued) The annealed Al-Mg alloys after 740 hours of alternate immersion stress corrsion exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn. Etched at 20 volts for 20 seconds in Barkers reagent. 500X.

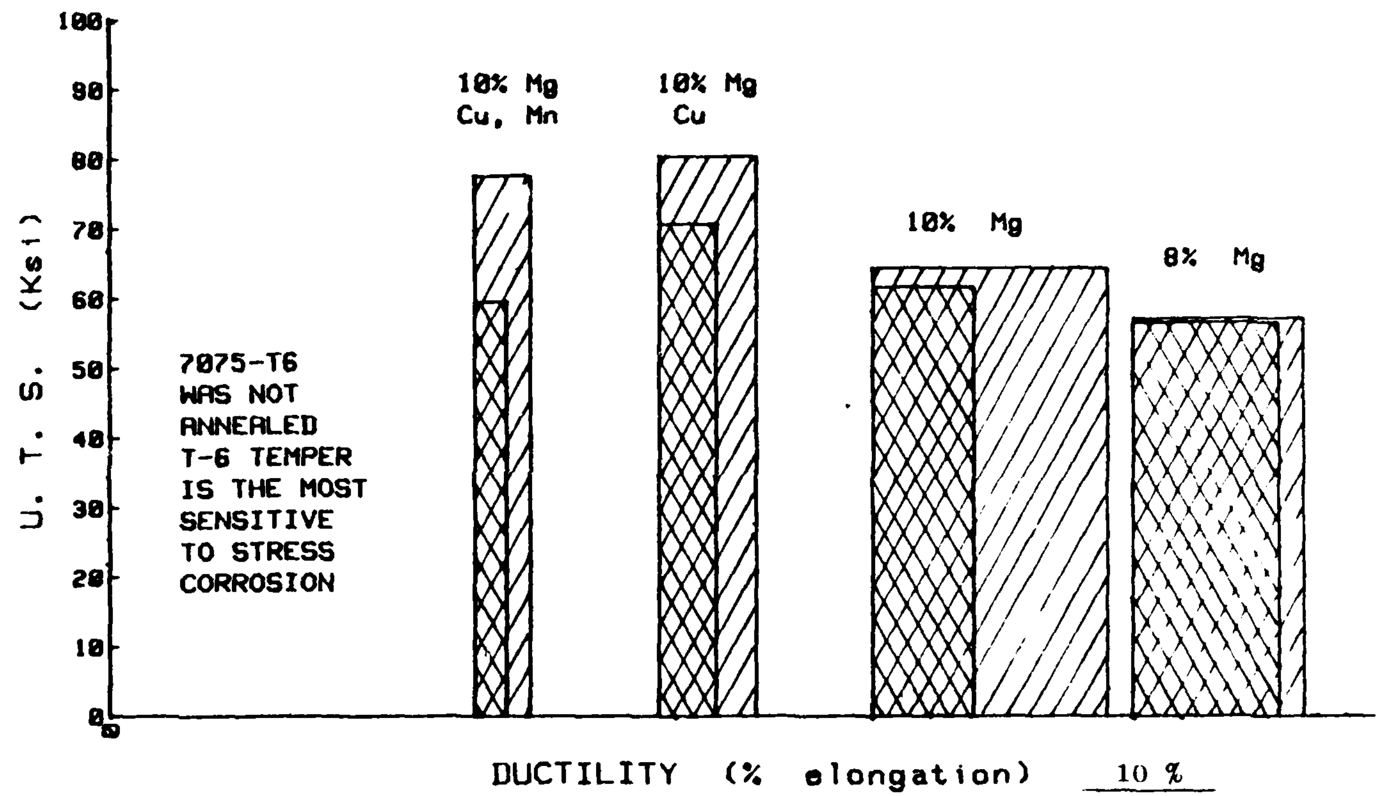


Figure 22. Comparison of mechanical properties of the warm rolled Al-Mg alloys in the annealed condition to the properties retained after stress corrosion exposure. The oxide was removed before exposure.

ANNEALED ▨ ANNEALED, POLISHED, STRESS CORRODED ▩

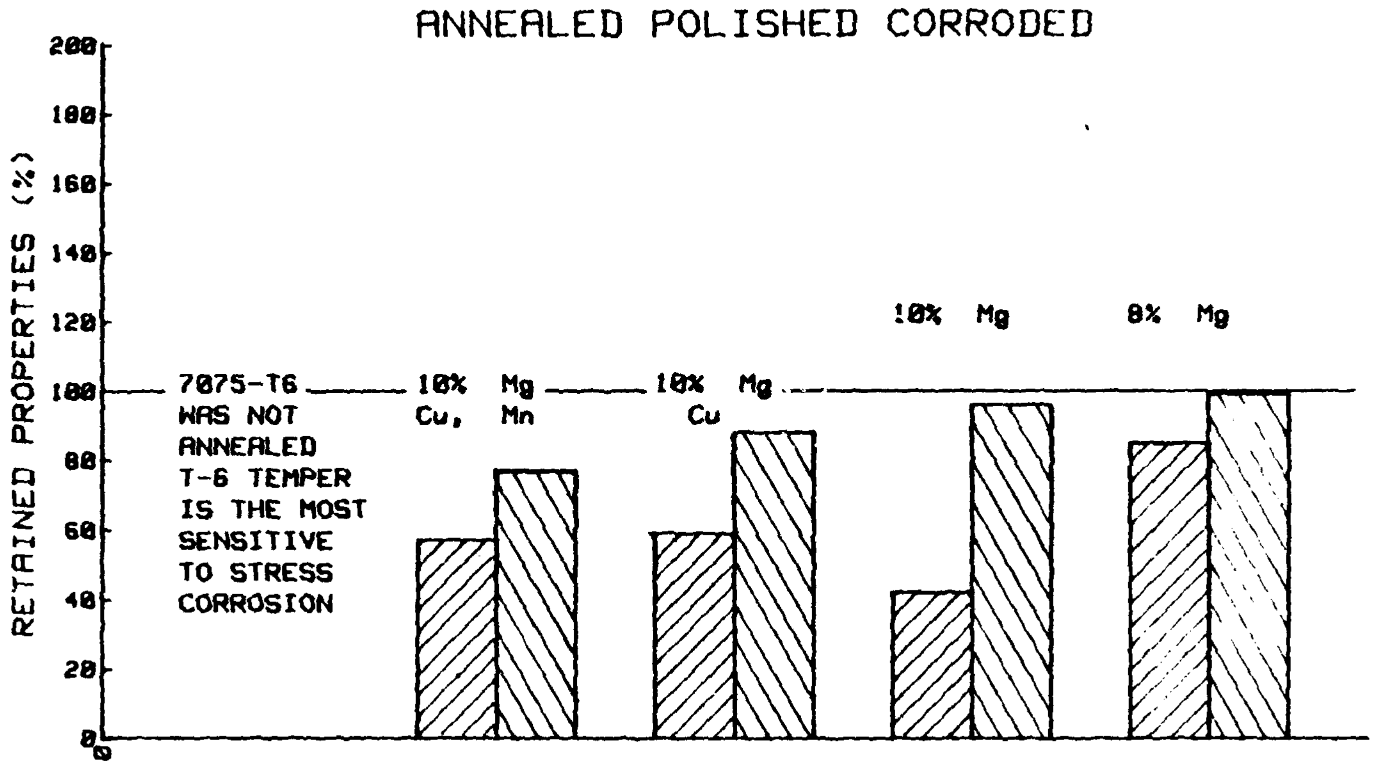


Figure 23. Effect of stress corrosion exposure on the annealed condition of the warm worked Al-Mg alloys. Here, data is represented as percentage of property retained, refered to the annealed condition before exposure. The oxide was removed prior to exposure.

DUCTILITY ULTIMATE TENSILE STRENGTH

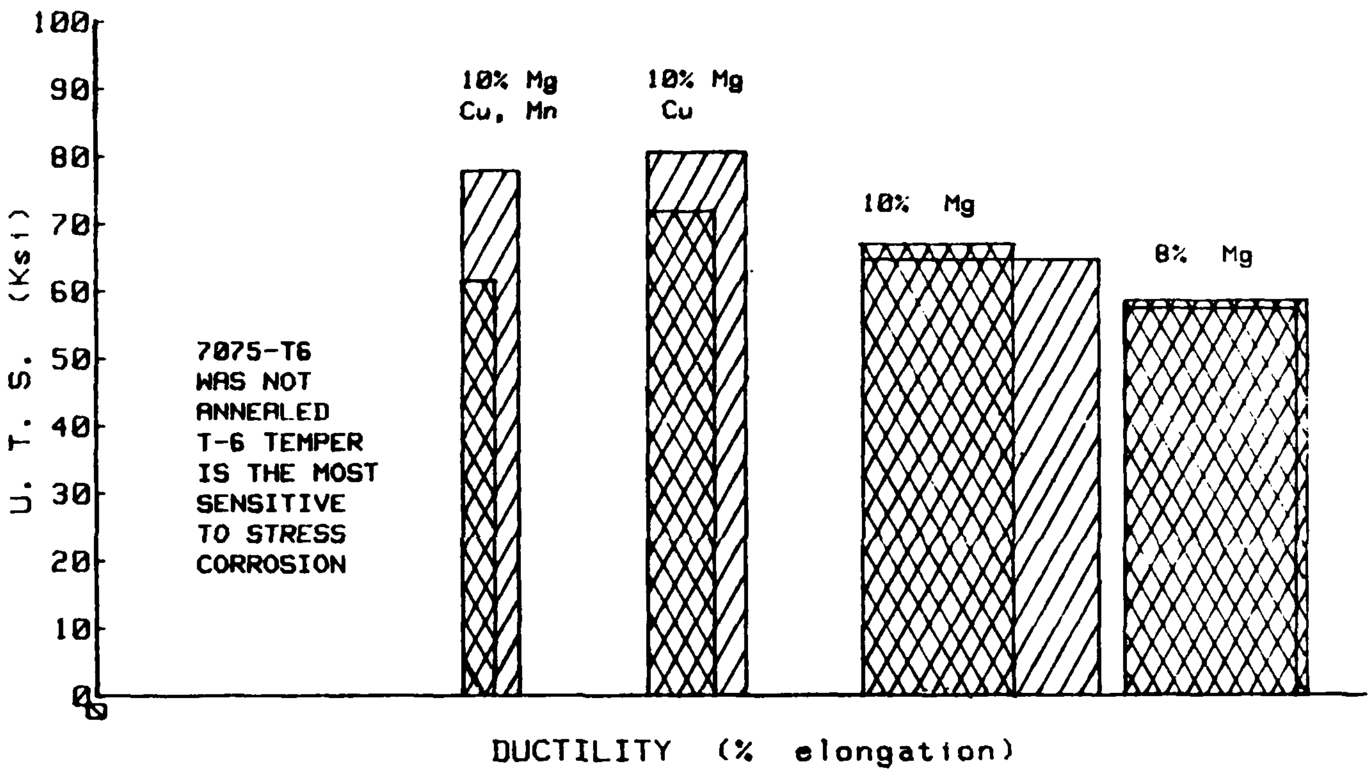


Figure 24. Comparison of mechanical properties of the warm rolled Al-Mg alloys in the annealed condition to the properties retained after stress corrosion exposure. The oxide resulting from annealing was not removed before exposure.

ANNEALED [single-hatched box] ANNEALED, UNPOLISHED, STRESS CORRODED [cross-hatched box]

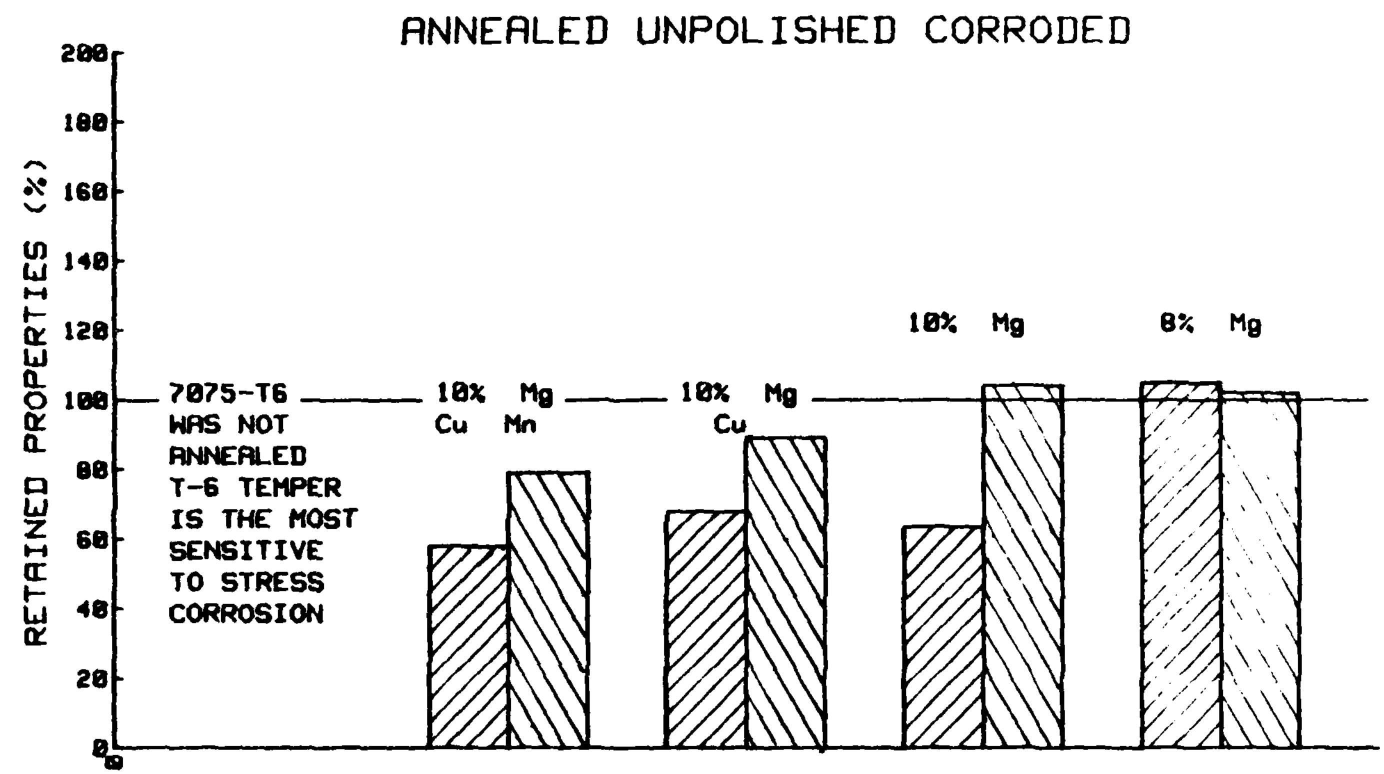
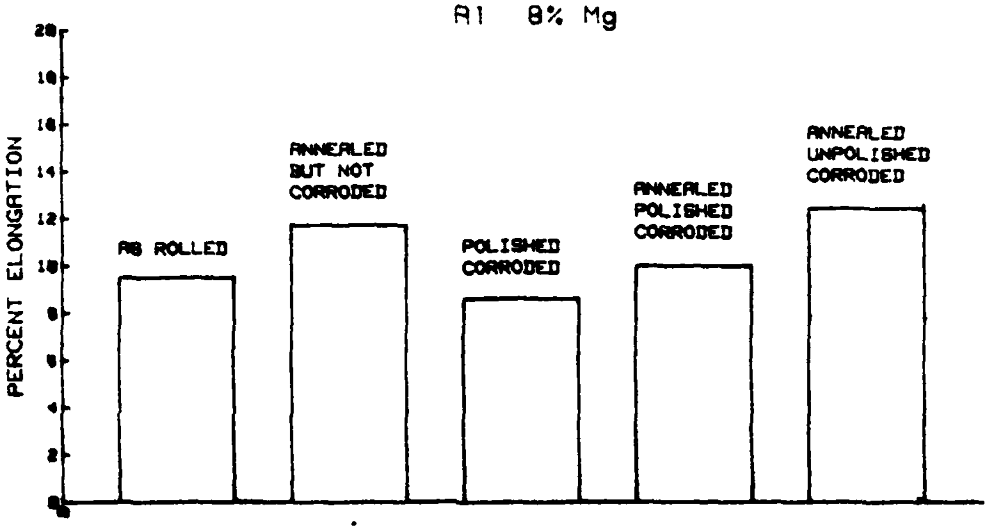
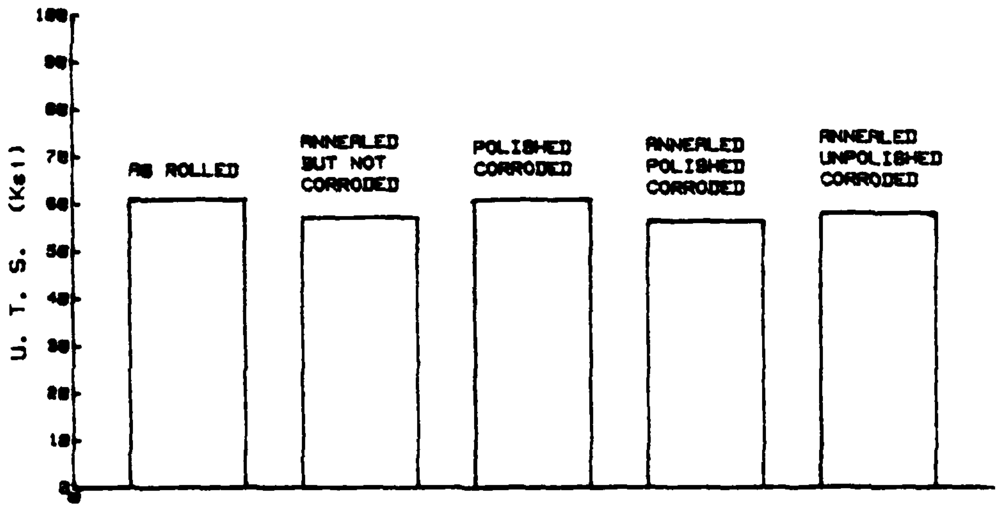


Figure 25. Effect of stress corrosion exposure of the annealed condition of the warm-worked Al-Mg alloys. Data is represented as percentage of property retained, refered to the annealed condition before exposure. The oxide was not removed before exposure.

DUCTILITY ULTIMATE TENSILE STRENGTH



(a)



(b)

Figure 26. Mechanical properties of the 8% Mg binary alloy expressed as (a) ductility in percent elongation (b) ultimate tensile strength (UTS) in the as-rolled condition, annealed condition, and following various stress corrosion exposures.

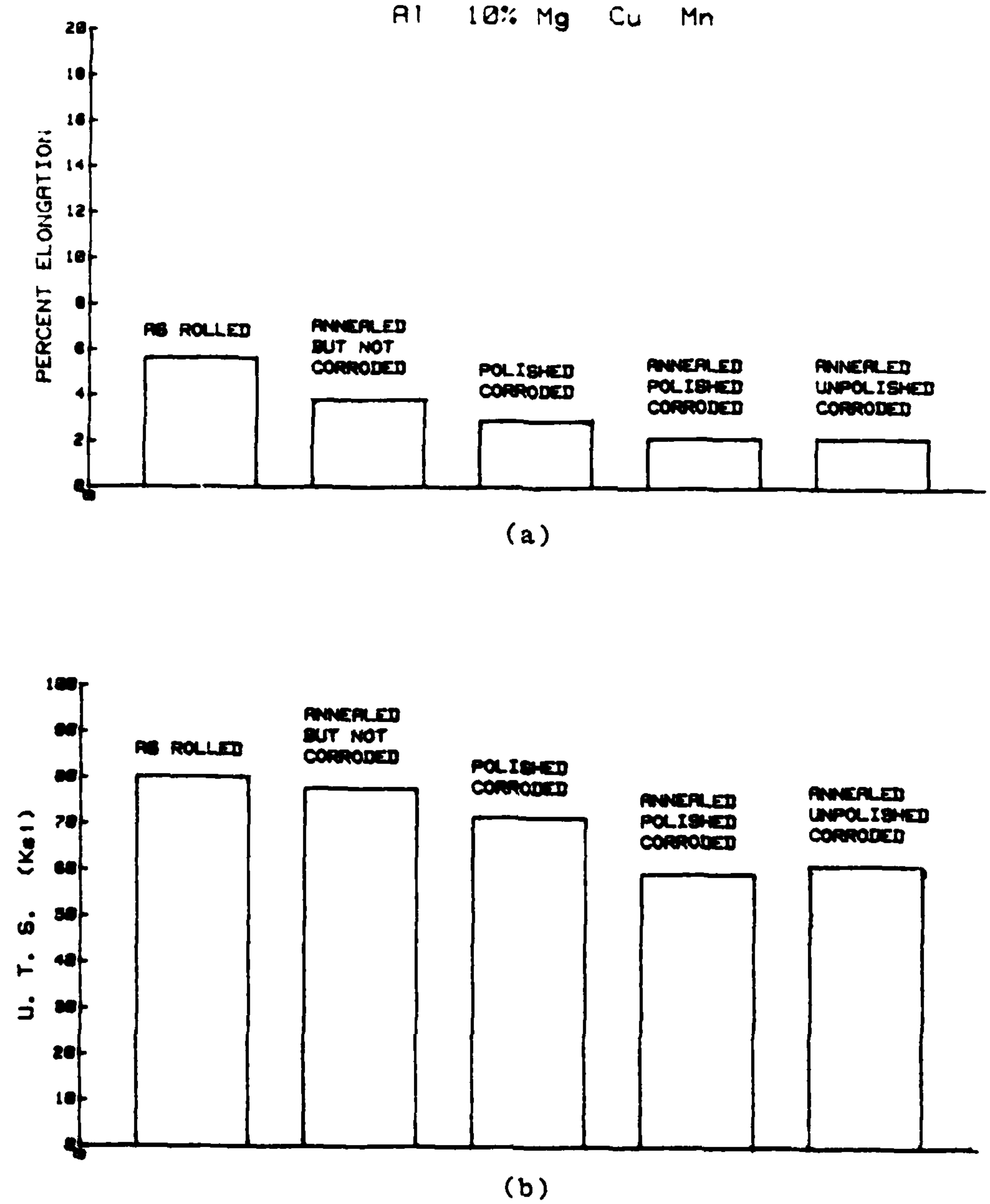
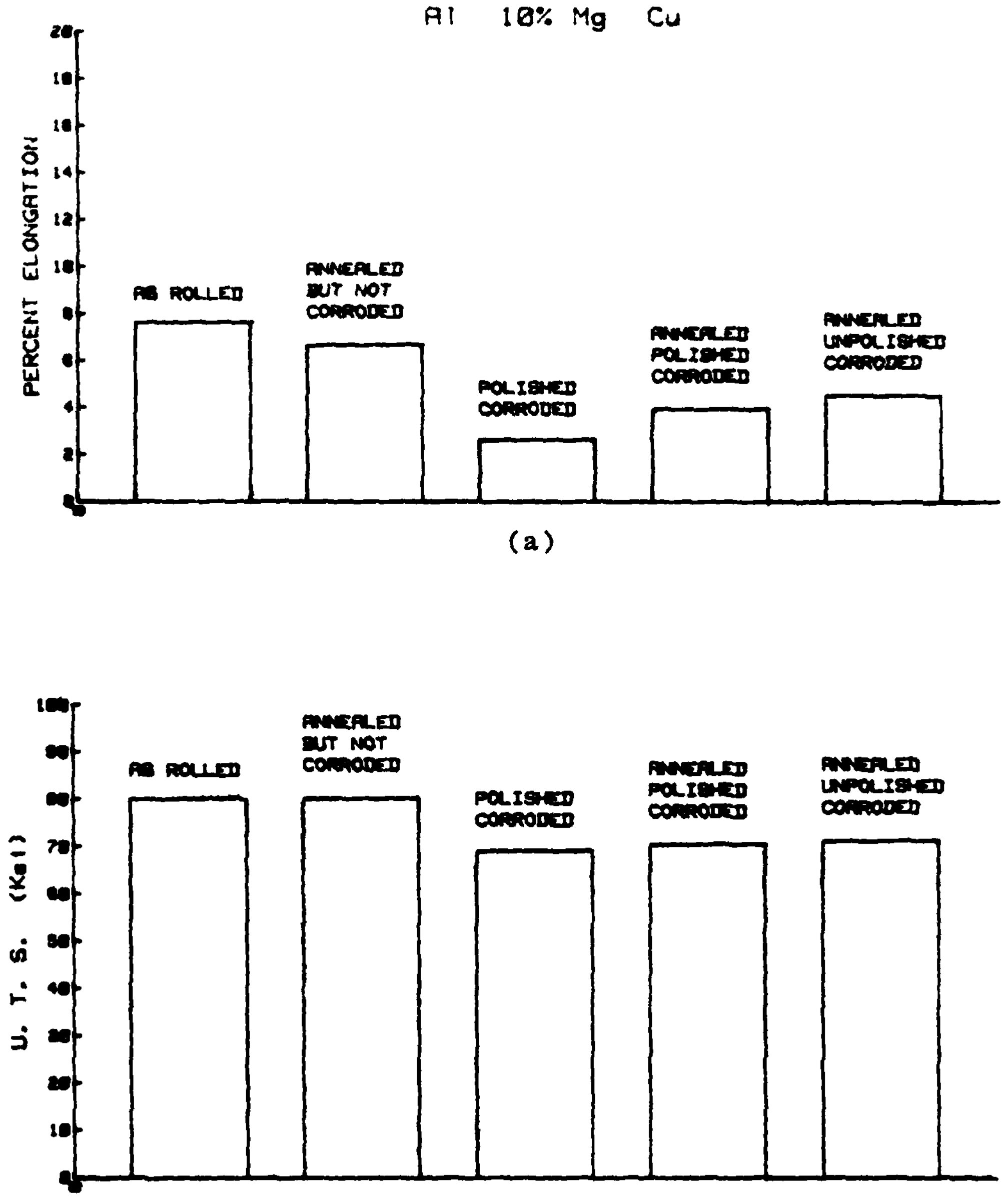


Figure 27. Mechanical properties of the 10% Mg, .43% Cu, .52% Mn expressed as (a) ductility in percent elongation and (b) ultimate tensile strength (UTS), in the as-rolled condition, as effected by the various stress corrosion exposures.



(a)

(b)

Figure 28. Mechanical properties of the 10% Mg, .43% Cu alloy expressed as (a) ductility in percent elongation (b) ultimate tensile strength (UTS) in the as-rolled condition, as effected by the anneal and as effected by the various stress corrosion exposures.

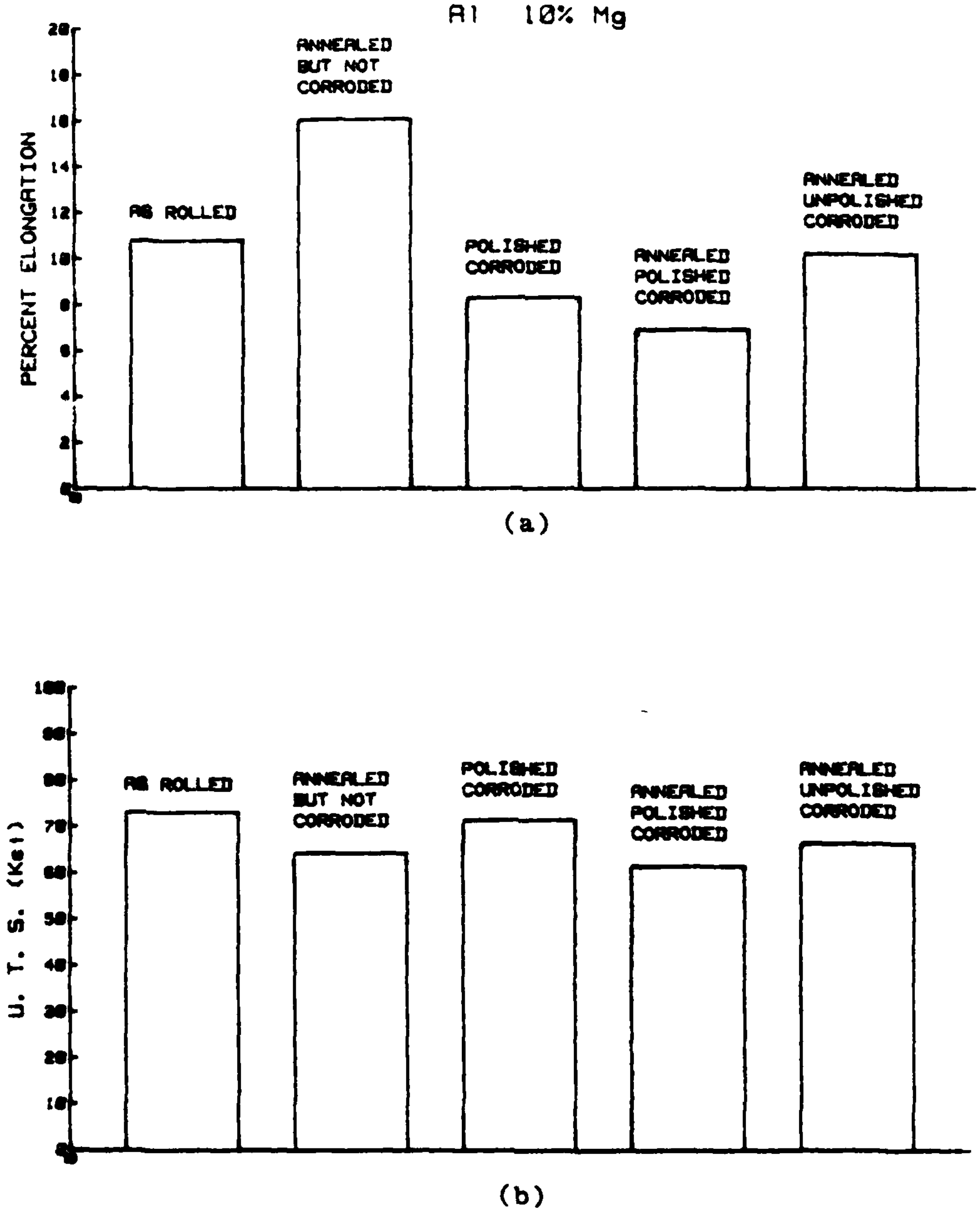
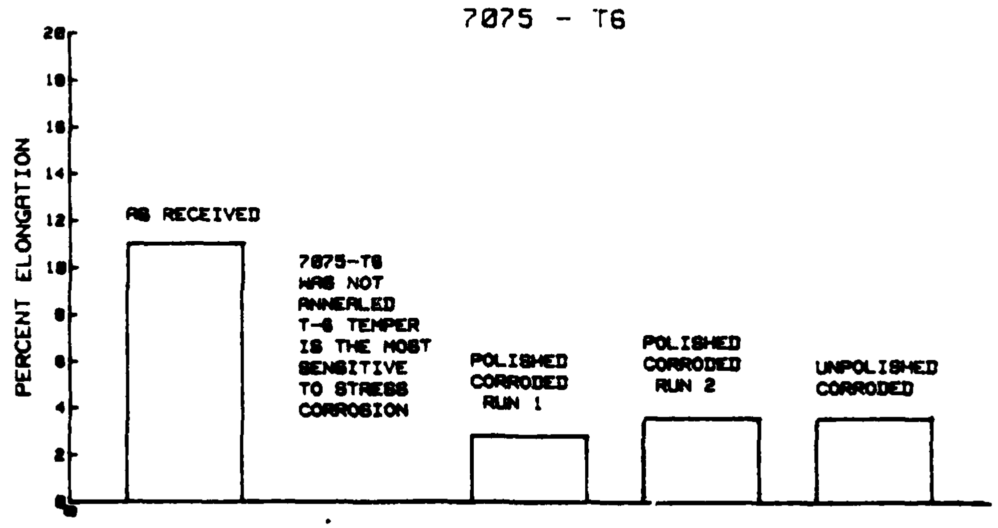
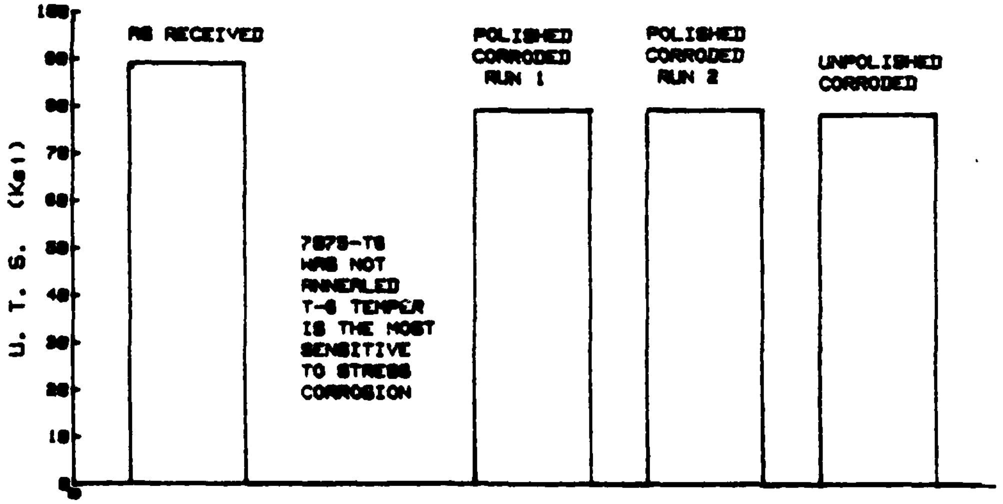


Figure 29. Mechanical properties of the 10% Mg binary alloy expressed as (a) ductility in percent elongation (b) ultimate tensile strength (UTS), in the as-rolled condition, as effected by the anneal and as effected by the various stress corrosion exposures.



(a)



(b)

Figure 30. Mechanical properties of 7075-T6 expressed as (a) ductility in percent elongation and (b) ultimate tensile strength (UTS), as received and after various stress corrosion exposures.

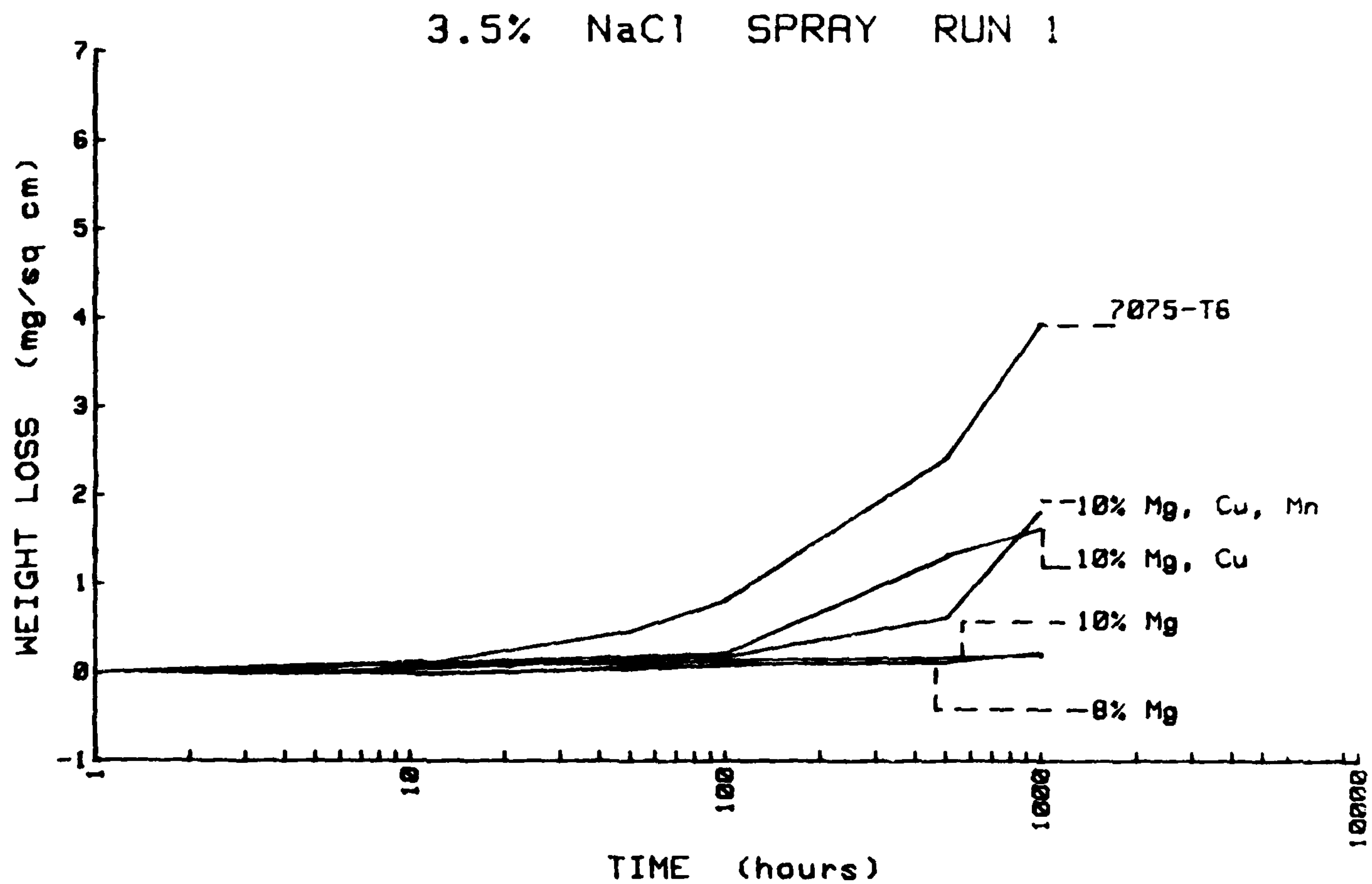


Figure 31. *Results of the first* accelerated general corrosion test, expressed as weight loss in mg per cm^2, versus time.

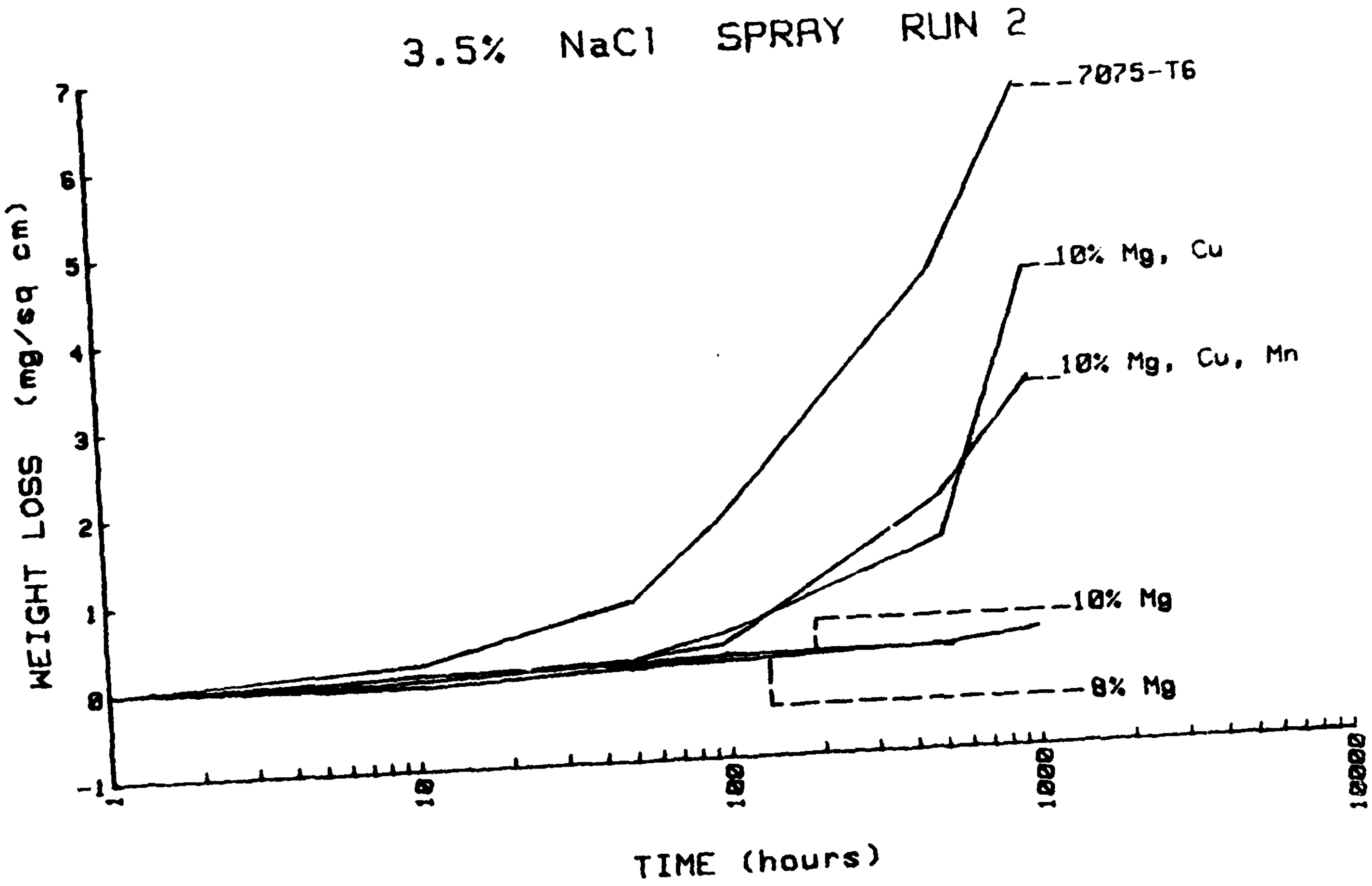
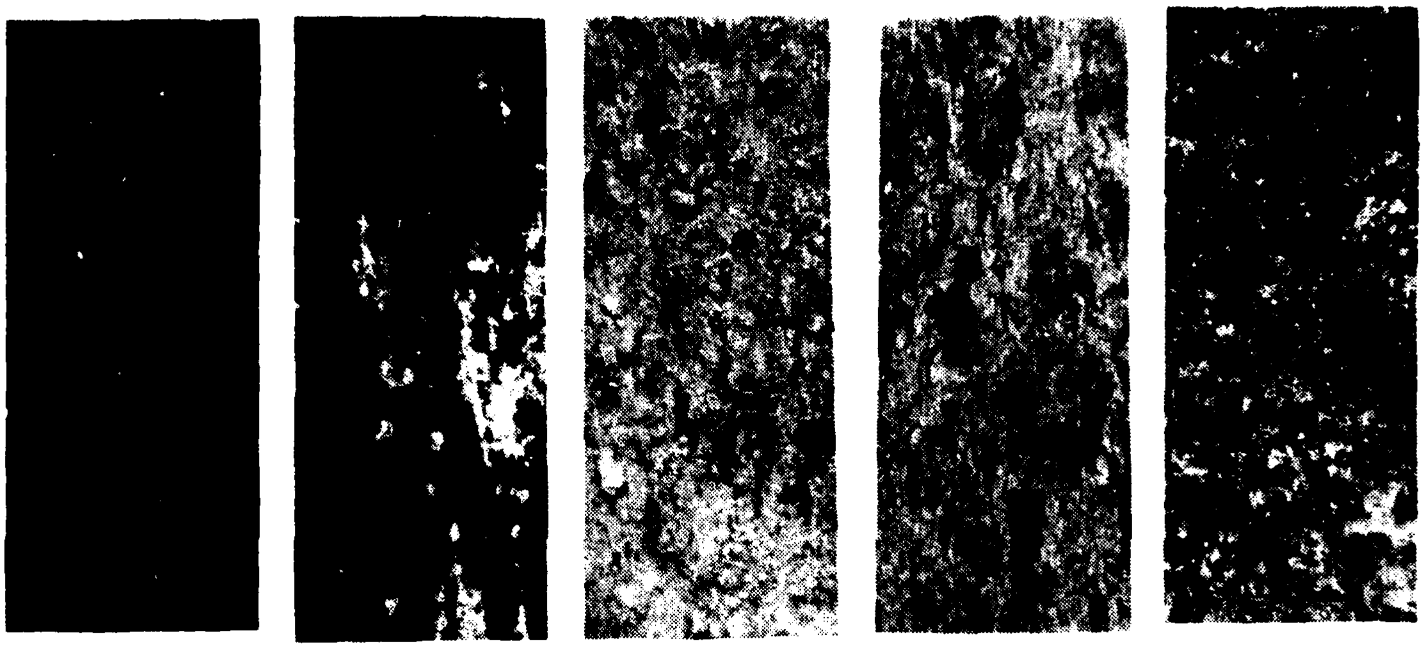


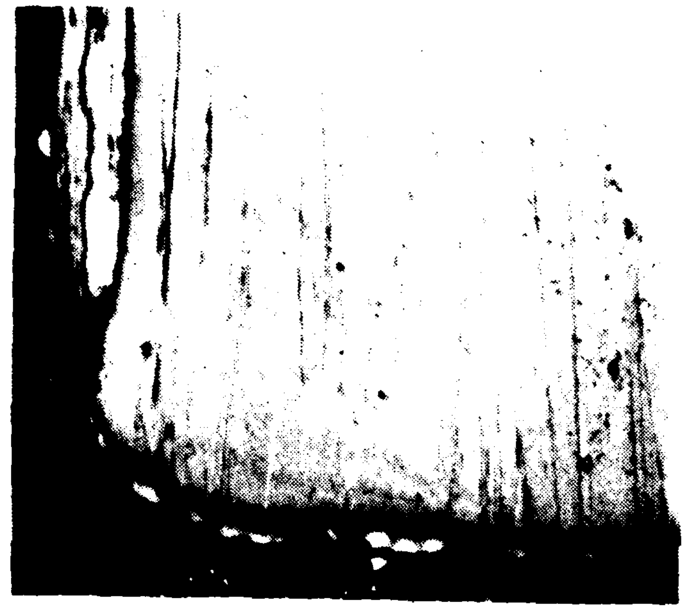
Figure 32. Results of the second accelerated general corrosion test, expressed as weight loss in mg per cm^2, versus time.



(a) (b) (c) (d) (e)

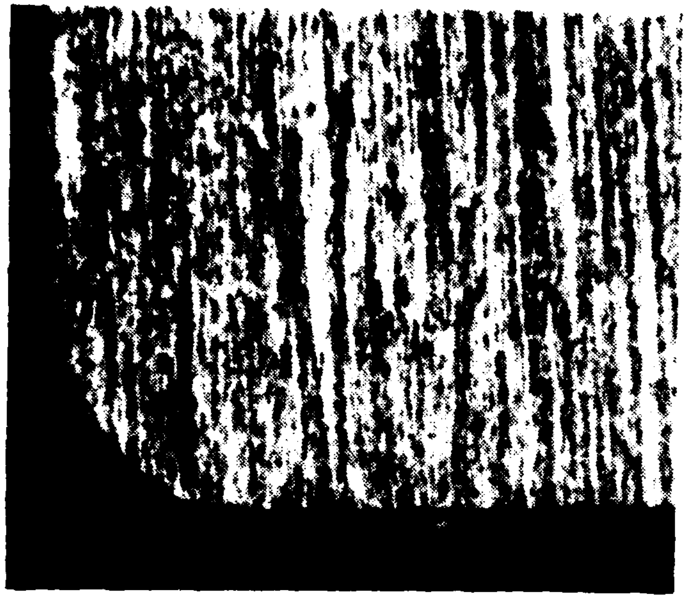
Figure 33. Macro photographs of the surface topography after 1000 hours of salt spray general corrosion exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6, 30X.

(a)

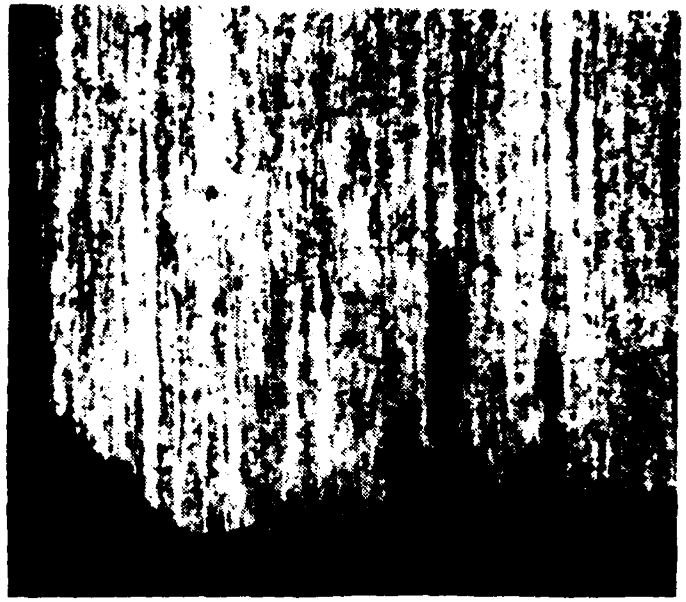


(b)

Figure 34. The ends of the general corrosion samples after 1000 hours of salt spray exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched at 20 volts for 20 seconds in Barkers reagent, 7075-T6 was etched in Kellers etch for 20 seconds. 500X.

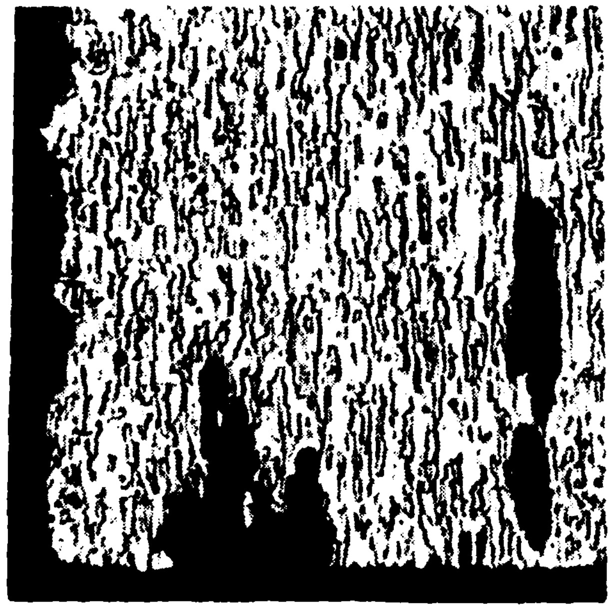


(c)



(d)

Figure 34. (continued) The ends of the general corrosion samples after 1000 hours of salt spray exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched at 20 volts for 20 seconds in Barkers reagent, 7075-T6 was etched in Kellers etch for 20 seconds. 500X



(e)

Figure 34. (continued) The ends of the general corrosion samples after 1000 hours of salt spray exposure: a) 8% Mg, b) 10% Mg, c) 10% Mg, 0.4% Cu, d) 10% Mg, 0.4% Cu, 0.5% Mn, e) 7075-T6. The Al-Mg alloys were etched at 20 volts for 20 seconds in Barkers reagent, 7075-T6 was etched for 20 seconds in Kellers etch. 500X

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